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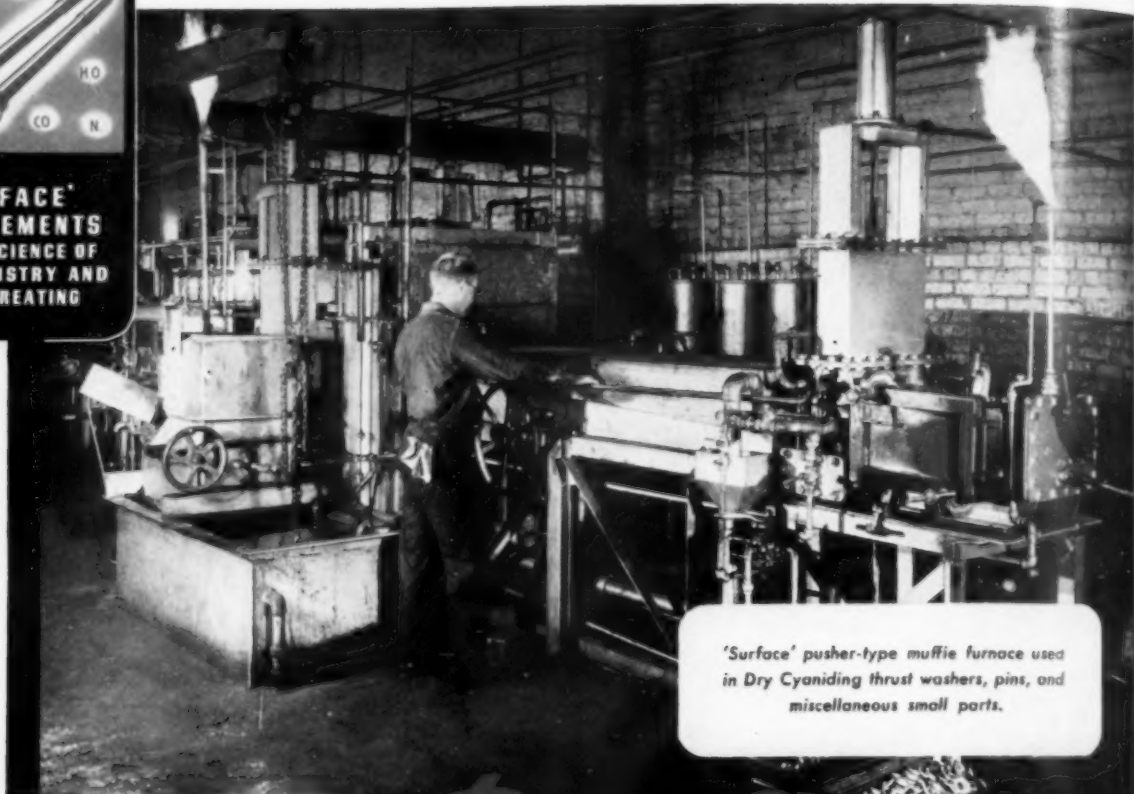
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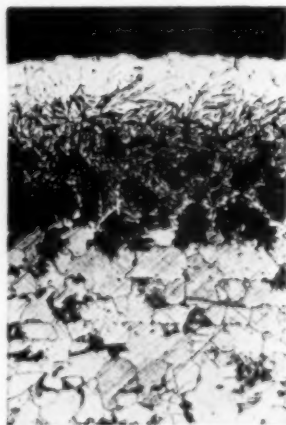


JUNE 194



'Surface' pusher-type muffle furnace used in Dry Cyaniding thrust washers, pins, and miscellaneous small parts.

ONE EXAMPLE:



SAE 1015

DRY CYANIDING

Photomicrograph shows a typical .006" dry cyanide case put on SAE 1015 steel by treatment for approximately 45 minutes at 1500° F. followed by slow cooling to room temperature.

Dry Cyaniding

combines features of both carburizing and nitriding with the advantage of flexible and continuous control of carbon and nitrogen in the case structure. Quenching is not a requisite for obtaining extremely hard surfaces by the dry (gas) cyaniding process. Work is perfectly clean and shows substantially no distortion due to heat treatment. Any carburizing or cyaniding grade of steel or iron may be treated by the dry (gas) cyaniding process . . . The process is another 'Surface' development to come from extensive research in the Science of Gas Chemistry and Heat Treatment. Write for Bulletin SC-124 for complete details.

'Surface'



*SPECIAL ATMOSPHERE, RADIANT-TUBE FURNACES FOR:
Gas Carburizing and Carbon Restoration (Skin Recovery), Clean and Bright Atmosphere Hardening, Bright Gas-Normalizing and Annealing, Dry (Gas) Cyaniding, Bright Super-Fast Gas Quenching, Atmosphere Malleableizing, Atmosphere Forging, and Specific Effects upon Metal Surfaces.

SURFACE COMBUSTION CORPORATION • TOLEDO 1, OHIO

METAL PROGRESS

Vol. 47

June, 1945

No. 6

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The front cover of this issue is a reproduction of a painting by Franklin Arbuckle showing the interior of the forge shop at Sorel Industries, Ltd., Sorel, Quebec, Canada, and kindly loaned by the Canadian firm.

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How Ryerson Assures Alloy Steel Quality

Each alloy shipment is personally inspected and tagged as a part of a rigid quality control system, which we call the Ryerson Certified Steel Plan. This plan covers: selection of the individual heats of alloy steel, the testing of samples from each heat, the positive identification of every alloy bar with stamped heat symbol and painted color markings (or with metal tags), and a Ryerson Alloy Steel Report, which is sent with each shipment. This report shows complete test information for the particular heat of steel used in filling your order. It contains: chemical analysis, recommended working temperatures, the Jominy hardenability results, and an interpretation of

physical properties for 1, 2, 3 and 4 inch rounds quenched and drawn at 1000°, 1100° and 1200° F. The report serves as positive identification, a check on quality and as a guide to satisfactory heat treatment.

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METALLURGICAL ASPECTS

OF ALLOY STEEL

AIRCRAFT ENGINE FORGINGS

TO THOSE who are familiar with aircraft power plants the title implies something special in materials and quality control. The proven dependability of engine and propeller assembly has been the result of manufacturing process controls applied continuously from the melting and rolling of the aircraft quality alloy steel through the forging, heat treating, machining, and assembly of the part. These careful control processes are carried even further. After assembly of the new engines they are given a "green run-in"—that is, they are run for varying periods after which they are completely torn down and all parts checked for evidence of unusual wear or galling. After re-assembly all engines are given a final run-in before shipment. One or two per cent are chosen by spot check for further dimensional checks and qualification testing. In addition, the engines and propeller assemblies are completely torn down after specified periods of service operation, thoroughly inspected by all known methods available, defective and worn parts replaced, then re-assembled and subjected to further proof checks to insure another period of trouble-free operation.

The chief difference between aircraft and automotive engine parts is in the design factors used, which in turn, influence material selection and processing methods. Aircraft parts are designed with low safety factors to secure lightness. This is combined with high induced stresses in high horsepower engines which are

continuously being increased in horsepower with only minor changes in the cross-sections of the highly stressed parts. For example, one of the early radial engines delivered 575 hp.; this same engine with only small changes in the alloy steel parts today delivers over 1350 hp.

Numerous failures of a certain part in an automotive engine may frequently be eliminated by redesigning the part to increase the critical section, thereby adding weight and rigidity. When a highly stressed aircraft engine part fails, loss of life may result, especially if a single-engine plane is involved. Generally, the designer is not at liberty to increase the weight and he, therefore, must take every advantage of the highest quality materials. The obvious answer to his problem is the use of alloy steel forgings made under rigid controls.

A high sense of engineering ethics exists between the steel producer, the forger, and the engine builder regarding aircraft steel specifications. This does not imply that deviations from

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specifications are impossible, but rather that a legitimate procedure has been set up to permit the use of material which deviates somewhat, if it will not affect the service of the part. Specifications were originally established on the basis that maximum uniformity from lot to lot was desirable and this uniformity could be attained by selecting metal, heat by heat, in the steel plant. Prior to the present war, heats that did not meet specification were rather easily diverted to other uses. During recent times, when these high quality electric steels were restricted mainly to aircraft, it became increasingly important that heats melted for aircraft purposes be so used. Thus, the metallurgist's role in time of war becomes increasingly important, as it is his duty and responsibility to avoid unnecessary rejections.

These same principles are also applicable to aircraft processing specifications. Frequently heat treating specifications have been written with a certain Brinell hardness range, but mechanical property requirements then have been applied which operate to restrict the Brinell limits by half. A few of these have been modified because the ranges were too narrow to be practicable, but a larger number have not and are being met only by precise metallurgical control.

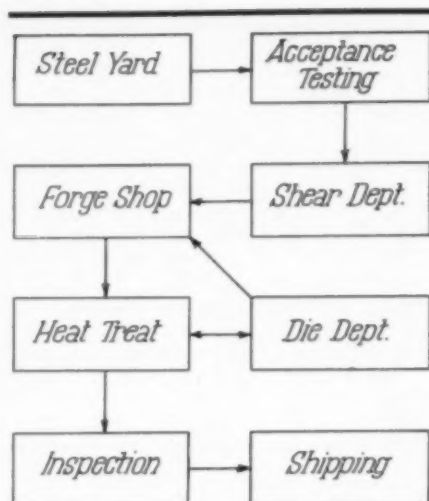


Fig. 1 — Flow Sheet of Forge Shop

Because material and process specifications are on such a rigid basis, salvage and deviation procedures are an integral part of aircraft manufacturing. The adequacy of these systems has been proven by the durability of the modern high horsepower engine.

To meet the above requirements the steels shown in the tabulation below are commonly used. Note the popularity of the 4340 type steel. Prior to 1937 there were six varieties of this chromium-nickel-molybdenum alloy steel — two Navy specifications, one Army, one Wright specification and a Pratt & Whitney specification as well

as the "standard" S.A.E. 4340 analysis. That year, during the SAE Convention, a conference was held at Atlantic City between representatives of the Army and Navy Air Forces, various melters of aircraft quality steel, forgers and engine builders, which agreed to concentrate on X-4340, now known as S.A.E. 4340. This Ni-Cr-Mo steel is one of the most reliable of the oil hardening types from the standpoint of uniform machinability, generally adequate hardness penetration, as well as comparative freedom from distortion.

The flow sheet of a modern aircraft forge shop is comparatively simple. It is similar to commercial forge shops; the chief difference lies in the rigid controls that must be established. The departments will be discussed in detail.

Steel Types Used for Aircraft Forgings

PART NAME	STEELS COMMONLY USED		
	OIL HARDENING MEDIUM CARBON	CARBURIZING	NITRIDING
Crankshafts	S.A.E. 4340	S.A.E. 2515, 3310	S.A.E. 4340 VCM (modified)
Propeller shafts	S.A.E. 4340	—	—
Connecting rods	S.A.E. 4340	—	—
Gears	S.A.E. 4340, 6150	S.A.E. 3310, 2515 A.M.S. 6260 S.A.E. 4620, 4820	Nitr alloy G (modified)
Cylinder barrels	NE 8740 (A.M.S. 6322) S.A.E. 4140	—	Nitr alloy G (modified)
Cams	—	S.A.E. 2515, 3310 A.M.S. 6260 S.A.E. 1020, 1117	—
Counterweights	—	—	—
Propeller blades	S.A.E. 4340, 4330	—	—
Propeller spiders	S.A.E. 4340	—	—
Propeller barrels	S.A.E. 4340	—	—
Propeller hubs	S.A.E. 4340	—	—

VCM: C, 0.28/0.38; Mn, 0.60/0.90; Cr, 0.90/1.20; Ni, 0.50/0.80; Mo, 0.90/1.10%

Steel Yard

Steel is received as bars or billets of various cross-section and in lengths from 10 to 30 ft. It is piled in bins with an electro-magnet crane and held until approved by the laboratory.

Each bar and billet 3 in. square or over has been individually identified at the mill by stamping the ingot number and the relative location from top to bottom in the ingot. (Only the bars adjacent to the top and bottom of the ingots are located in sizes from 1½ in. diameter to 3-in. round-cornered

Multiple

Tong Hold

Upset

Blocked



Final Hammer Operation

Trimmed

Flange Upset

Fig. 2 — Forging Sequence for Three-Way Spider in Propeller Hub Assembly.
 Steel: A.M.S. 6415. Weight of forging: 80 lb. All photographs by R. Stranieri

squares. Heat numbers are, however, stamped on every bar.) Smaller bars are bundled and tagged with the heat number only.

Proper bar identification is fundamental; without it, the acceptance tests that have been planned to insure quality, speed production and reduce rejections are worthless.

Laboratory Acceptance Testing

Before any heat of steel is released for cutting, it must be completely tested to determine conformance to specifications. The various acceptance tests which may be applied to aircraft quality steel include (a) macro-etch, (b) cleanliness, (c) chemical analysis, (d) hardenability,

(e) grain size and (f) mechanical properties.

Macro-Etch, by immersing cross sectional disks in hot 1-1 hydrochloric acid, when properly interpreted, is one of the most useful tests for detecting gross defects such as seams, porosity or pipe. Disks are cut from bars or billets corresponding to the top and bottom location of the first, middle and last ingots. If the heat of steel is to be used on a critically stressed part, macro-etches may be cut from bars or billets corresponding to the top and bottom location of every other ingot, or even every ingot in the heat. If the macro-etches are not satisfactory, additional discards are taken and new etch tests are made.

Cleanliness Rating—The micro-cleanliness method of checking non-metallic inclusions has

been, in the past, the standard method. With the advent of magnetic particle inspection of finish machined parts, it was noted that the micro-cleanliness test was not indicating the true magnaflux cleanliness of the forging steel. It was therefore necessary to develop a method of checking cleanliness of the stock which would indicate whether or not finished parts would pass magnaflux standards when the forgings were subsequently machined. The Wyman-Gordon Co. pioneered in the development of the magnaflux method of checking the cleanliness of aircraft quality steel and today it is the standard used by all steel mill suppliers and most of the users of

tered sections may be rough turned, or forged to rounds and subsequently rough turned. All samples are then heat treated in the same manner as the finished forgings and suitably ground or polished. Generally, the wet continuous method of magnaflux testing is used.

Chemical Analyses—Complete chemical analyses are made on drillings taken at half-radius from test prolongations adjacent to the bottom of the first ingot and the top of the last ingot. Additional carbon checks are made from random top and bottom locations. Other analyses are for manganese, phosphorus, sulphur, chromium, nickel, molybdenum, silicon and vanadium.

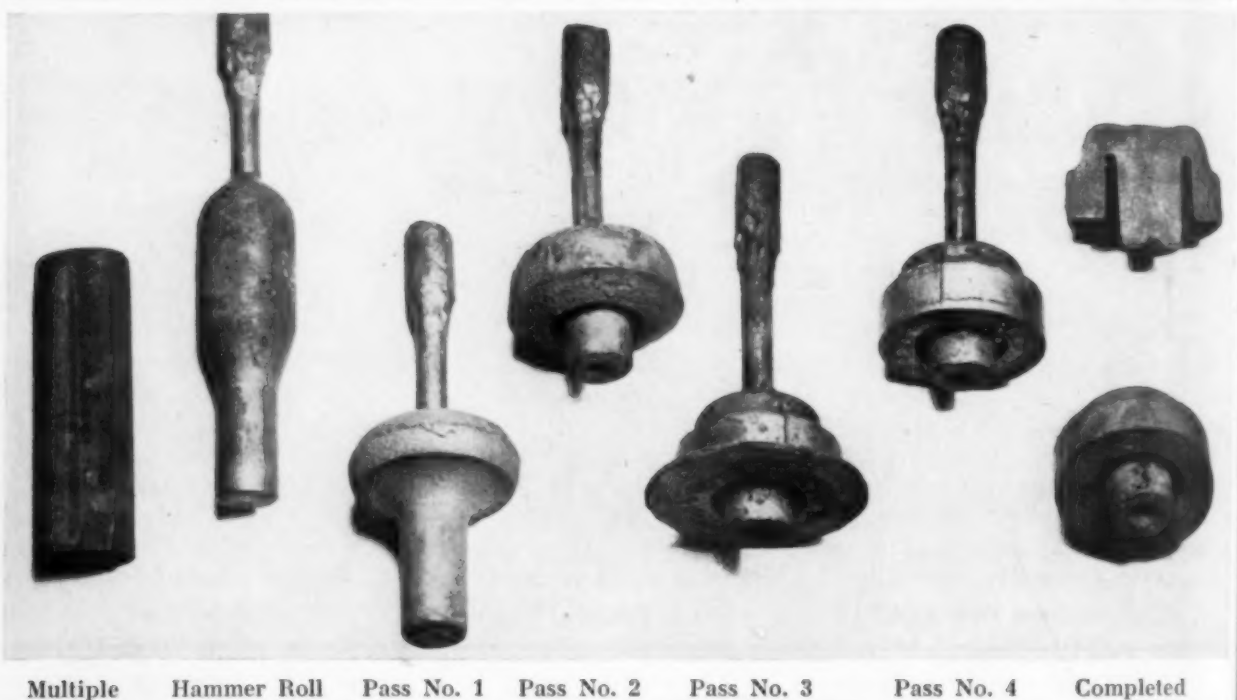


Fig. 3 — Reduction Gear Pinion Produced by Combination of Hammer Forging and Upsetting. Steel: A.M.S. 6250. Weight of forging: 6.2 lb. Note depth of groove, as shown by sectioned forging at upper right

aircraft quality steel. (See A.S.T.M. Standard E45-42T, and the topic "Magnaflux Testing" in S.A.E. Handbook, 1944.)

Samples for this test are generally taken from the bottom of the first ingot and the top of the last ingot. For highly stressed parts such as crankshafts, master rods and propeller shafts, additional locations may be tested.

For certain types of forgings, such as master and articulated rods, where the finish machined surface in various sections corresponds to the original center section of the bar or billet, the so-called quadrant method is used. The quar-

Hardenability—Two types of tests are required by aircraft specifications. The A.M.S.* hardenability test consists of oil quenching steel disks representing the average sections of the engine parts. The disks are first copper plated, heated for a specified length of time, quenched in oil and the hardness across the face of the disk, corresponding to one-half radius, is measured in Rockwell C-scale.

Recently there has been a tendency on the

*Do not confuse A.M.S. with the A.S.M.! A.M.S. refers to the Aeronautical Material Specifications of the Society of Automotive Engineers.

part of a number of manufacturers to specify the Jominy end-quench hardenability bar in addition to or in lieu of A.M.S. hardenability. In either case, samples are taken from the same locations as the turnings for chemical analyses, for experience has shown that these are most apt to show the maximum spread in hardness. The test has been standardized in 1943 S.A.E. *Handbook*.

Grain Size is generally specified as No. 5 or finer, according to A.S.T.M. specification E19-39T, Method (a), with grains as large as No. 3 permissible. Certain applications of alloy carburizing grades for crankshafts require a grain size of No. 2 to 5 to aid in making steel as clean as possible, but these heats are all fine-grained at normal quenching temperatures. It is becoming increasingly common to measure the grain size after the steel has been processed according to the production heat treat cycle instead of after carburizing at 1700° F. for 8 hr.—the Method (a) mentioned above. This is especially true of the medium carbon grades where Vilella's etch or some other suitable method may be used to show the grain size.

Mechanical Properties—The test coupons cut from each heat may be forged to the bearing diameter of the crankshaft or propeller shaft, to a diameter equivalent to the maximum section of the forging for which the steel is intended, or to some other suitable diameter. These forged test coupons are then heat treated as specified for the forgings for which the heat will be used, after which the surface hardness is checked by the Brinell machine. Tensile test pieces (and Izod impact test pieces, if specified) are cut from these coupons, and the results used to predict details of the production heat treatment on the forged parts as well as serving to evaluate suitability for the parts intended.

The method of centering tensile tests varies considerably. Some engine manufacturers still specify a point midway between the surface and center of the coupon—or half-radius, as it is familiarly known. Inasmuch as this position on forgings for cranks and propeller shafts is generally completely removed in machining, other manufacturers specify that the tensiles be centered two-thirds of the distance from center to surface, or $\frac{3}{8}$ to $\frac{1}{2}$ in. in from the surface at time of heat treatment. Data from the latter locations more nearly indicate the actual tensile strength of the material in the area adjacent to the finish machined surface.

Typical properties specified by aircraft manufacturers are as shown in the table above. (Yield strength is at 0.2% offset.)

Specified Physicals for Aircraft Forgings

YIELD STRENGTH	TENSILE STRENGTH	ELONGATION	R.A.	IZOD IMPACT	BRINELL
100,000 psi.	125,000	17.0%	55.0%		262/302
115,000	135,000	16.0	50.0	45	286/321
120,000	140,000	16.0	50.0	35	302/340
130,000	150,000	15.0	48.0		311/352
140,000	160,000	14.0	40.0		331/364
150,000	170,000	11.0	35.0		352/415
170,000	185,000	10.0	30.0		375/430

Shear Department

After material has been approved by the laboratory, the bars or billets are cut into forging multiples of the correct length and weight. 4340, 6150, 6135 and nitralloy are sheared hot—that is, at approximately 400° to 1000° F., in sizes up to $5\frac{3}{4}$ -in. round-cornered squares. Material is sheared cold up to $3\frac{1}{2}$ in. square for steel such as 2515, 4340 and 3310. Billets larger than $5\frac{3}{4}$ -in. square are cut by power hack saws or circular saws. Abrasive cut-off saws are used generally for small sizes, and can be used for billets up to the maximum capacity of the saw. Double multiples are sometimes torch-cut on each end and then cold sawed in the center, and then used with the torch-cut on the long hold end. In some instances both ends of the multiples are torch-cut, after preheating.

Shear-cut multiples are inspected on the ends for flakes, which show up as slight but unmistakable tears on the sheared surface. Saw-cut and torch-cut ends are magnafluxed, using prod contacts; flakes appear as ragged, irregular indications adjacent to the center.

Multiples are racked, and the skids or tote boxes are identified by metal tags bearing the heat number and the heat code number which subsequently will be stamped on the forging.

Die Department

Hammer forging dies are made of 0.50% carbon alloy steel (chromium-molybdenum) heat treated to several ranges from about 300 to 444 Brinell. The harder blocks are ordinarily used for shallow impressions and the soft blocks for deep impressions in dies for 12,000-lb. hammers or larger. Typical analyses quoted in *Metals Handbook* and *Metal Progress Data Sheets* are:

	HAMMER DIES		HOT WORK
	A	B	STEEL
Carbon	0.45/0.55	0.50/0.60	0.35
Manganese	0.60/0.80	0.65/0.85	0.40
Silicon	—	—	1.00
Chromium	0.75/1.00	0.80/1.10	5.00
Molybdenum	0.25/0.45	0.30/0.50	1.75
Vanadium	—	0.05/0.15	—
Tungsten			1.25

Mechanical press and hydraulic press dies are made from either of the above types. If made from the Cr-Mo-W hot work steel they are heat treated to about C-45 to 55. Certain hydraulic press piercing tools are made of alloy cast iron; it is suitable where high temperatures and pressures cause other die materials to weld to the steel being forged.

Trimmer dies are frequently made from rolled plate or forged plate normalized and drawn, and then stellite and ground on the cutting edges.

Forge Shop

The forge shop is, of course, the heart of the plant. The trend in aircraft forging is toward the use of heating equipment with complete temperature control. A good heater can make an excellent quality forging if he has only a box furnace as a heating medium; however, under war-time conditions and with inexperienced help, the advantages of the continuous furnace are apparent. Continuous furnaces are generally of the rotary hearth type, and vary in diameter from 5 to 20 ft. Some pusher type furnaces with refractory hearths are used, 20 to 30 ft. long. Induction heating as well as salt bath is used for special jobs. The fuels are conventional —

oil, gas, and electricity. Furnaces are usually controlled at 2350° F., with the actual temperature of the multiple itself varying from 2100 to 2250° F., depending on the type of steel.

The most popular piece of forging equipment is the steam drop hammer which varies in size from about 1000 lb. up to 50,000. Board drop hammers are used in sizes from 500 up to 6000 lb. Horizontal upsetting machines are found in sizes from 1 to 9 in. Hydraulic presses are of the water, oil or steam type. Fast acting mechanical presses are also used. Forging rolls may be used for pre-forming operations prior to final forging. Accessory equipment includes tong hold presses, trimming presses and crank pin twisting machines for crankshafts.

The usual forging steps include blocking, fullering, rolling and edging operations which rough form the product, then forging in the finish dies. The flash is trimmed (and the pins on crankshafts are twisted into position). If the finished forging conforms to the finish hammer die, aircraft forgings are generally given one final blow in this die to correct any distortion from trimming. Larger size forgings of the higher alloyed steels may be cooled in pits or furnaces to prevent thermal ruptures. Photographs of representative forgings, step by step, are shown on adjoining pages.

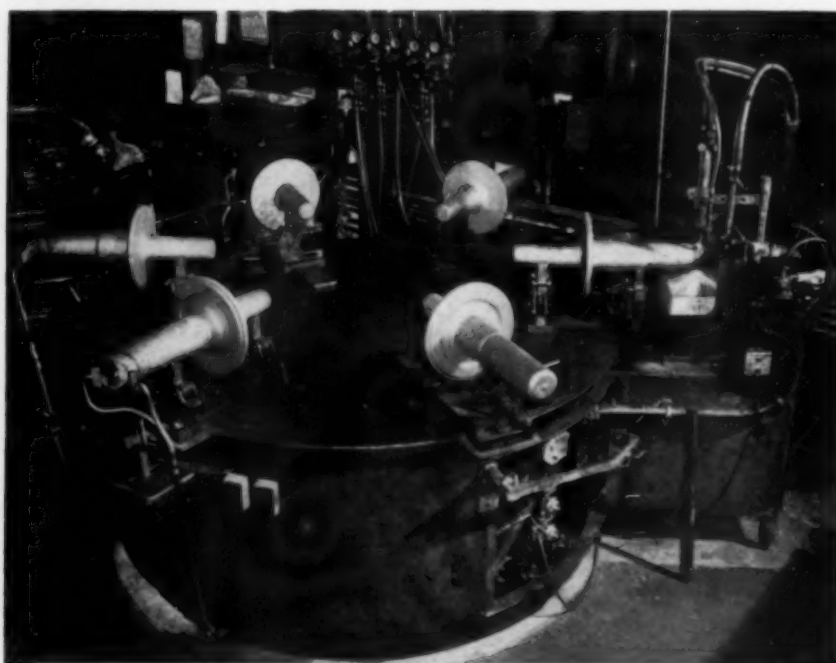


Fig. 4 — A Merry-Go-Round With Six Stations, Used for Preparing a Lug on End of a Forging, and Flame Hardening and Tempering the Lug Prior to Fracture Test

Heat Treating Shop

Heat treating equipment found in a modern aircraft forge shop generally consists of continuous furnaces, although batch types have been used successfully, especially for smaller parts. The continuous furnaces are usually pushers or rotaries; the fuel is gas, oil, or electricity. Some furnaces are equipped with controlled atmosphere, and the majority of the draw furnaces, recently installed, are of the recirculating type. Most heat treated aircraft forgings are oil quenched, although some are quenched in water, or in air, and one prominent manufacturer uses a caustic quench. Types of quenching equipment used are the flight conveyor for small parts, the automatic elevator type in which the tray is

quenched with the work, and the ferris wheel quench for larger parts. Gleason or Gogan quenching presses are used for some parts.

Forgings are either heat treated for machinability or are given a complete heat treatment for specified strength and ductility. Carburizing steels such as 2515 may be normalized and drawn, normalized, or normalized, quenched and drawn for machinability. S.A.E. 3310 is generally normalized and drawn, since cycle annealing requires too long a time at holding temperature to insure complete transformation. S.A.E. 4620 steel may be normalized or cycle annealed depending upon the machining which will follow. Carburizing steels are subsequently rough machined, copper plated where stop-off is necessary, carburized and then hardened and drawn in the conventional manner.

The medium carbon alloy steels such as S.A.E. 4340, 4140, 3140, 4640 and NE8740 may be normalized and drawn, but recently have been cycle annealed at constant temperature to improve their machinability. The cycle used on 4340 steel consists of austenitizing at 1700° F. for 1½ hr., cooling to 1200° F. and holding there from 5 to 8 hr. A number of "S-curves", considered useful in determining annealing cycles, have been published recently by the U. S. Steel Corp.'s research laboratory and the Crucible Steel Co. of America's research laboratory.

Inspection and Testing

Forgings are cleaned by shot blasting, sand blasting or pickling, then inspected visually for laps, mis-match or under-fills. Gages and fixtures are used to insure proper centering and indexing, and odd shaped parts may be individually checked on special jigs. Magnaflux aids visual examination where necessary.

Grain flow is examined on etched slices and is modified if necessary by die changes or other means. The macrograph alongside shows the grain flow of a large crankshaft for an in-line, liquid cooled engine.

Each heat treated forging is Brinelled in one or more places and sections are regularly cut for hardness penetration and tensile tests. The hardness-penetration tests are then correlated with the Jominy end-quenched tests for the heat, and the combined data are used to predict the response of other heats.

Some parts, such as crankshafts, propeller shafts, master rods, spiders and barrels, are forged with a fracture test coupon attached. This integral coupon is designed primarily to detect overheating. However, an experienced and careful inspector can observe differences in the texture of the fractures that result from the omission of any one of the heat treating operations. A difference in grain size

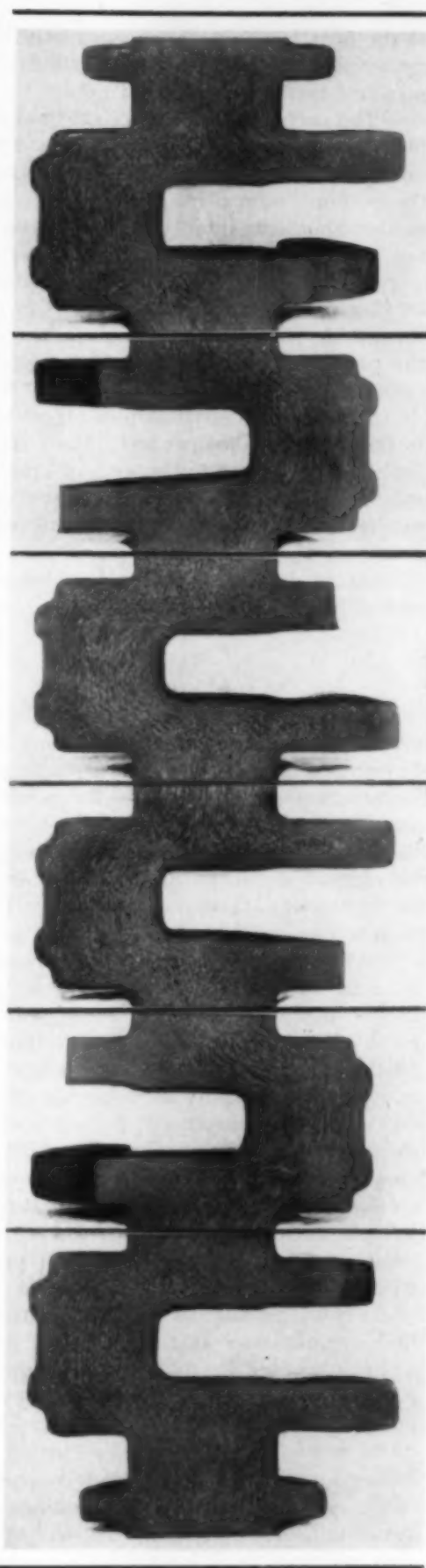


Fig. 5 — Six Throws of a Crankshaft for an In-Line Engine Were Sectioned on Center Line and Deep Etched to Show Grain Flow. Overall length of actual shaft was 42 in. Courtesy of Western Division, Wyman-Gordon Co. Laboratory

at the heat treating temperature (say No. 4 to 5 rather than No. 7 to 8) and certain intangibles manifest themselves on the fracture.

The surface Brinell of a heat treated forging, plus the Jominy test bar readings and the hardness penetration data for that heat, together with the extremely sensitive fracture test, give information about the quality of the forging that cannot be duplicated by any other mechanical tests — *provided* a careful study of the heating and forging operations has been made to determine the correct size and location of the test piece, and the proper amount of work to be done on it in each of the forging dies.

The fracture test occupies a position similar to that of the magnaflux test. Both are powerful tools for the improvement of quality; both require skilled, intelligent interpretation; but neither is a substitute for sound engineering practice. Figure 4, page 1112, shows one method of cutting, flame hardening and tempering one end of a forging for fracture test.

Salvage

As with all manufacturing plants, salvage is an integral part of the forge shop operations. Forgings that do not conform to internal shop standards are diverted from the production line as soon as they are found and are channeled to the various salvage departments, where they may be rechecked for hardness and re-heat treated as indicated. Outsized parts may be ground, re-struck or locally forged.

All parts are Brinelled that are outside the internal shop limits, including those parts that may conform to blueprint requirements. The establishment of limits closer than those required by the customer is sometimes necessary, because many of the generally used specifications, as now written, have inconsistent Brinell and tensile or Brinell and Izod requirements. This inconsistency becomes greater as the hardenability of the steel diminishes, with the result that the internal Brinell limits frequently must be varied from heat to heat if other requirements are to be met without excessive re-heat treatment.

From the salvage Brinell department, forgings are returned to production if satisfactory, or are re-routed for additional salvage work.

One of the most common salvage operations is the re-draw. If the forgings have been found to be hard, they are re-processed at a slightly higher draw temperature and tested for hardness. Again, it is the low hardenability steels — low compared to the section size to be heat treated — that are responsible for much of this re-treatment.

Test data secured as outlined under the heading "Acceptance Testing", together with production data, are used as the basis for establishing the original draw temperature for the heat. If the hardenability spread is large for tests taken from various locations in the heat, then the draw temperature that has been set to make sure that the low hardenability parts of the heat are within hardness specification is too low for the balance of the heat, with the result that part of the lot is too hard.

If the hardenability of the heat as a whole is low, then the surface hardness necessary to produce specified mechanical properties may have to be restricted to the top half of the Brinell range, which means that a relatively large number of the forgings may be over the top limit on the first draw and, therefore, must be re-drawn.

Re-hardening is necessary whenever forgings are below the surface hardness known to be necessary to meet certain mechanical properties, even though that hardness is within the specified range. In addition to these are the few soft forgings that are found in production. Forgings to be rehardened are usually sent through the regular production hardening unit and then re-drawn at a temperature slightly lower than that used originally.

Visual and magnaflux inspection often show laps, wrinkles or other defects that are frequently superficial. Their depth is checked by grinding until they are removed; the forging is then inspected to see if there is sufficient metal remaining to clean up when machined.

Re-striking is used when forgings are oversize because of worn dies, are mis-matched because of shifting dies, or have bent flanges or other light sections that cannot be readily straightened immediately after normalizing or other heat treatment.

Oversize forgings are re-forged in new dies at the start of a new run and the excess metal is forced out into the flash.

Mis-matched forgings are re-struck after the dies have been adjusted, and bent forgings are given one light hammer blow to re-align them. All these operations are closely supervised with re-striking temperatures lower than the original forging temperature.

Local forging is used on forgings that are undersize or bent at certain small areas or on forgings that have been ground close to finish to remove surface imperfections. It has the obvious advantage of avoiding the necessity of reheating the entire forging (as for re-striking) and has the further advantage of refining the areas re-worked at the low finishing temperatures.

TROUBLE SHOOTING

PROBLEMS IN

STEEL MACHINING

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IN MANY PLANTS the departmental foreman is left to work his own way out of machining problems that arise. Realizing the added burden that such a procedure places on the already overworked supervisor, other companies refer machining puzzles to an individual or a department that can more properly cope with the various phases of the troubles as they appear.

During the past few years the writer has been called upon for this purpose many times, and gradually has worked out a useful method of approach. The steps about to be explained are listed in general order of their value. The reader must realize, however, that on many jobs the cause is apparent, or the trouble is of a recognizable and familiar type. In such cases many of the preliminary steps to be given are unnecessary. Generally speaking, if the investigator follows a definite routine, just as a chemist does in analyzing an unknown, it will result in a quicker and more accurate solution of the problem.

Trouble usually shows up in one, two or three ways. First as wear, chipping or breakage of tools; second as loss of size or poor finish on work; and third as machine trouble such as break-downs or power loss. The first object, then, of the "trouble shooter" should be to determine the *type* of trouble. Next he should obtain and tabulate all data regarding the job, including all obvious facts such as feed, speed, tool material, tool condition, tool hardness, material being

machined, hardness, machine condition, and so on, and a number of other facts such as the history of the job, source of material, and previous performance of tools, machine, and operator. Once this information is obtained the job is ready to be analyzed, realizing, when considering each point, that all of the other variables are assumed, at first, to be guiltless. Later the variables can be combined to arrive at the total answer.

Step I — Cutting Tools

(a) *Material* — Are the cutting tools made of the correct material?

Each kind of tool material can do only certain types of work. When reckless substitution is made trouble usually results. For example, on one job a drill "burned up" when first put into use; the extremely bad condition of the drill suggested some trouble in the tool. A spark test showed that the drill was made of carbon tool-steel instead of high speed steel as specified. The speed run, correct for high speed steel, was much too high for carbon steel.

Four general types of material are used for cutting tools. (This statement appears elementary. Many of the statements will appear elementary. Much trouble shooting applies elementary common sense.) Carbon toolsteel is used where speed and feed are low, or where conditions are such that generate low heat from abrasion. High

speed steel is most commonly used for machining on production jobs, and finds its chief application where medium speeds and feeds are used. Its advantage over carbon steel is its red hardness, or ability to retain hardness at temperatures approaching 1000° F. Cast non-ferrous metals with inherent hardness and wear resistance, such as stellite, are used for more severe conditions than high speed will stand. These tools, however, are not as tough as high speed and will chip more easily. For highest speeds and heaviest feeds, tools with cemented carbide inserts are used. The brittleness and cost of these tools limit their use somewhat, but they can be run at rates far exceeding those possible with other materials.

More will be said about these four materials when feeds and speeds are discussed.

(b) *Hardness* — Is the tool of proper hardness for the job?

Too soft a condition will cause excessive wear. Too hard a tool may chip or break. Usually Rockwell C-62 to 63 hardness should be used on carbon and high speed steel where a tough tool is required, while hardness as high as C-65 to 66 can be used for greater wear resistance. Quite often wear resistance must be sacrificed for greater toughness. Sometimes cobalt high speed steel can be used to increase red hardness and wear resistance.

High hardness can be reduced on carbon and high speed steels by re-tempering. Low hardness can be increased only by a complete new heat treatment. Under step I (d) methods of increasing surface hardness will be taken up.

(c) *Proper Heat Treatment* — Is an incorrect heat treatment the cause of the trouble?

Too high a temperature in hardening may easily coarsen the grain structure, resulting in brittleness, cracks, and warpage. Too low a temperature (non-solution of carbides) causes softness, spotty hardness, spalling and low wear resistance. Some improperly treated tools can be salvaged by re-treating; in most cases, however, an overheated tool will never be satisfactory. Improper tempering should be watched for. A hard, brittle tool will result from too low a tempering temperature or too short a time, while tempering at too high a temperature will reduce the hardness too much. Very often the re-tempering of a properly tempered high speed cutter will considerably improve its toughness and general performance. For example, a run of gear tooth rounding cutters chipped very badly in service, but after re-tempering at 1100° F. they performed satisfactorily.

(d) *Surface Condition* — Has the treatment produced a soft skin?

Frequently a soft, decarburized surface is found on tools after heat treatment. This "decarb" is more likely on molybdenum high speeds than on the old-fashioned 18-4-1 tungsten high speed steels, and is due as much to loss of molybdenum as to carbon. Soft skin is especially prevalent on tool bar stock when it is received from the mill and all too often the tool is made so that the cutting edge is right at this surface. Such a tool will not stand up in service. The "decarb" or "bark" should always be removed

Table I—Suggested Rakes and Clearances

	HIGH SPEED STEEL	CAST ALLOY TOOLS	CEMENTED CARBIDES
Side clearance	8 to 10°	7 to 9°	6°
Front clearance	8 to 10°	7°	6°
Back rake	8 to 12°	5 to 10°	0
Side rake	12 to 15°	5 to 15°	3°

before forming a tool from bar stock, or before grinding the tool after heat treating. Many tool steel manufacturers recommend that $\frac{1}{32}$ in. be removed from all surfaces before making a tool from the bar.

Greater wear resistance can often be obtained on high speed tools by nitriding the finished tool by heating it in a mixture of sodium cyanide and potassium cyanide for 10 to 30 min. at about 1000° F. If the tool is re-ground all over, the nitriding process must be repeated after each grind; if only one face is ground, the nitriding is done only once.

Increased wear resistance and lower surface friction can often be obtained by a chromium "flash"—electroplating the cutting edge in a standard chromium plating solution for 1 to 2 min. This, of course, must be done after each grind. In one case where a blanking punch was a little too soft and "picked up" during service, its life was increased from 15 pieces per grind to 3000 by chromium plating.

It should be noted here that nitriding and chromium flashing are used to *improve* tool life, and not as a general substitute for proper material or treatment.

(e) *Correct Size* — Continual breakage of tools may be a sign that the tool is too heavily loaded. In such a case the only solution may be to increase the size or strength of the tool, or to reduce the feed of the machine.

(f) *Correct Grinding* — A careful watch should be made for edges burned in grinding and for grinding checks where cutting edges do not stand up. Overheating of tools during grinding, or plunging of heated tools into cold water to

cool them off, causes checks and cracks. When surface checks or burned edges are found it is best to remove them by further grinding. Considerable improvement may at times be had if those surfaces which rub against the chip are smoothly honed.

(g) *Correct Rake Angles* — Improper clearances or too great a rake angle may cause excessive wear or weakness at the cutting edge. Specific angles for cutting tools are given in the literature of many tool manufacturers. General recommendations are listed in Table I. Experience of the foreman is of great value here. Proper "chip breakers" and "curlers" must also be considered when long chips are fouling up the work.

(h) *Proper Setting in the Machine* — Here again the experience of the foreman is of great value. Particular care should be taken to avoid too much overhang or interference of tools. All tools should be checked for rigid clamping.

Step II — Material Being Cut

(a) *Surface Condition* — Poor surface condition is often the cause of excessive tool wear. Particular offenders are parts that have to be

cut in such a manner that the tool cannot get below the surface immediately. Broaching of scaled holes, form turning, and drilling into scaled surfaces are examples of this. Forging scale or scale remaining after heat treatment is hard and abrasive, and chips and wears the cutting edges. A glazed surface sometimes causes trouble in gripping for some types of machining. Molding sand on the surface of castings is a most troublesome thing. Cold rolling or cold straightening will work harden the bar, and sometimes the effect at the surface will be large enough to impair its machining.

An unusual case occurred recently when machining a low carbon steel forging. Excessive surface wear was found to be caused by a high carbon surface, picked up during a pack annealing operation. The cure was to reduce the temperature of the heat treatment to below A_{c1} , the lower critical, below which carbon is not absorbed by steel. Scale or sand can be removed by shot-blasting. Heavy scale may have to be pickled. Shot-blasting will also roughen a glazed surface to facilitate gripping. When the surface has been heavily cold worked or glazed a draw at 1100 to 1200° F. will help.

Trouble Chart for Machining Operations

TYPE OF TROUBLE	SYMPTOMS	REMEDY (ALL OTHER FACTORS BEING PROPER)
1. Tools		
(a) Material	Excessive wear, chipping, breakage	Change to harder or more wear resistant material Change to tougher material
(b) Hardness	Excessive wear, chipping or breakage	Soft tool — re-harden or change tool Hard tool — re-draw tool
(c) Heat treatment	Spalling, brittleness, wear, spotty hardness	Overheated tool — re-harden or change tool Underheated tool — re-harden or change tool
(d) Surface condition	Excessive wear	Re-grind surface of tool; nitride or chrome flash
(e) Size	Continual breakage	Select larger size tool, or use less feed
(f) Grinding	Breakage or chipping	Re-grind or select new tool
(g) Cutting angles	Excessive wear; breakdown of cutting edge	Adjust angles
(h) Proper setting	Chatter or breakage; wear at one side	Too much overhang; move tool in or support it Rubbing work; reset tool
2. Material being cut		
(a) Surface	Wear or chipping of tools	Pickle, shot blast or draw steel
(b) Hardness	Gummy, tearing cuts Excessive wear	Re-condition steel or increase speed of cutting Re-condition steel; re-draw steel or reduce speed
(c) Inclusions	Excessive wear or chipping	Throw out steel or reduce speed and feed
3. Feed and speed	Rough cutting and finish Tool wear Tool wear and breakage	Increase speed Decrease speed Decrease feed
4. Coolant	Pick-up on tools Wear on tools Heating and smoking Rough finish Chipping and wear on tools	Increase sulphur compounding Decrease compounding; thin oil if possible Decrease compounding, thin oil, use soluble oil Increase compounding of oil Increase or adjust coolant flow

(b) *Hardness*—Selection of proper hardness for the material to be cut is sometimes a "King Solomon" problem. When a soft part drills all right the broach operators complain of tearing. If the hardness is raised to accommodate the threaders, the turning department starts to have trouble. Best general hardness for steel has been found to be about as follows:

Plain carbon steel (0.10 to 0.30% C)—130 to 170 Brinell
 Alloy steel (0.10 to 0.30% C)—150 to 180 Brinell
 Plain carbon steel (0.35 to 1.00% C)—160 to 210 Brinell
 Alloy steel (0.35 to 1.00% C)—180 to 220 Brinell

Best results in broaching, tapping, threading, shaping and gear tooth cutting are found when the hardness runs near the high limit. For drilling, milling, sawing and turning, the lower side of the hardness range gives better results. Too low a hardness will result in poor finish, tearing and inability to hold size, while too high a hardness causes excessive tool wear.

The microstructure of steel has also been found to play an important part in good machining. Lamellar or slightly spheroidized pearlite gives the best results on steels with less than 0.60% carbon. Medium or coarse-grained steel will machine more freely than fine-grained steel.

For unannealed steel on which the microstructure is satisfactory, or for annealed steel on which the hardness is too high, a high draw for 2 to 4 hr. at 1150 to 1300° F. will usually improve machinability. Decreasing speed and feed on hard materials will also reduce tool wear. When the steel is too soft for good machining and a re-anneal or normalizing treatment is not practicable, an increase in speed will often help the finish of the part.

(c) *Inclusions*—Non-metallic slag inclusions, either in heavy slug form or finely divided, are a serious source of trouble. The symptom is excessive tool wear or chipping. There is usually not much that can be done to remedy this situation except to scrap the steel or to struggle through the trouble at reduced speed and feed. (These remarks, of course, do not apply to inclusions of manganese sulphide or lead particles

which are intentionally added by the steel maker to improve machinability and embrittle the chip for screw machine operations.)

Step III — Speed and Feed

Usually light feed and high speed, or heavy feed and low speed are considered to go together. General recommendations regarding speed and feed are found very liberally distributed in manufacturers' literature. The book entitled "Manual on Cutting of Metals With Single-Point Lathe Tools", issued in 1939 by the American Society of Mechanical Engineers, can also be consulted with profit.

In fixing speeds and feeds the material used for tools is obviously a major consideration. Suggested rates are given in Table II. The general rule is: Low speed and light feed for carbon tools; medium speed and medium feed for high speed tools, high speed and medium or medium-heavy feed for cast tools; and high speed and heavy feed for carbide tools. Low speeds and feeds on carbide tools (and often on cast tools) usually result in poorer tool life than 18-4-1 high speed tools will give.

Very often the size or shape of work determines the speeds. Large, heavy castings must be turned at very low speeds because of the difficulty and high power requirements of running at speeds which would be recommended for small parts of the same material. Very thin parts must be turned at very light feeds to prevent bending of the part under the tool load.

Generally speaking, the harder the steel the lower the speed of cutting. For example, the speed of turning a medium carbon alloy steel drops from 90 ft. per min. to 50 ft. per min. as the hardness increases from 200 Brinell to 250 Brinell. At the same time tool life is shortened.

Step IV — Coolant

A liberal quantity of the proper coolant is a very essential factor for proper machining. The function of the coolant is (a) to cool the tool, work and chip, (b) to lubricate the point of the cutting tool, (c) to prevent pick-up or welding of the chip to the tool, and (d) to help carry away chips and lubricate moving parts near the tools. A liberal quantity is considered to be from 2 to 5 gal. per min. per cutting tool. The stream should be adjusted so as to hit the point between tool and work, or so that it can be carried to that point. Care should be taken to prevent the escaping chip from splashing the

Table II — Suggested Speeds for Turning (Ft. per Min.)

MATERIAL	HIGH SPEED STEEL	CAST ALLOY TOOLS	CEMENTED CARBIDES
Cast iron	50 to 100	100 to 150	200 to 400
Low carbon steel (screw stock)	120 to 150	150 to 200	400 to 500
Low carbon steel	90 to 130	130 to 180	300 to 400
Medium carbon steel (soft)	70 to 100	90 to 140	200 to 300
Medium carbon steel (semi-hard)	50 to 80	80 to 120	150 to 250

coolant everywhere but on the work.

The proper selection of coolant may spell the difference between success and failure in machining. Sulphur or chlorine compounded mineral oils should be used for soft, gummy steels. The compounding prevents welding of the chip to the tool and promotes cleaner cutting. As the material becomes harder an oil with less compounding should be used, and more cooling value introduced by thinning the coolant as much as possible (using water for soluble oils and paraffin oil for regular oils). Addition of lard or other fatty oils will aid matters where additional lubricity is needed.

Several examples will serve to illustrate these remarks: A regular sulphur-compounded soluble oil was used for general high speed turning. When a job requiring carbide tools was introduced, this same oil was used. The carbide tool chipped and did not last long. The oil was changed to a very thin mixture of a non-compounded soluble oil. Tool life increased 300%.

A lightly compounded mineral oil was used for shaping gear teeth. When a run of soft, gummy metal appeared, the chips welded to the cutters. The oil was changed to a heavily compounded mineral oil and all traces of pick-up disappeared.

During the hard hobbing of shafts at 400 Brinell hardness, one lot of shafts came through slightly high in hardness. The hobs dulled after only two pieces. The oil was changed from a medium compounded mineral oil to an uncompounded paraffin oil. Hob life immediately went back to normal.

Step V — The Machine

When trouble appears in the machine, such as lack of rigidity, loose bearings, or wrong camming, the job is for the machine repair department. The foreman generally can handle this



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We, upon whom has been placed the responsibility of leading the American forces, appeal to you with all possible earnestness to invest in War Bonds to the fullest extent of your capacity.

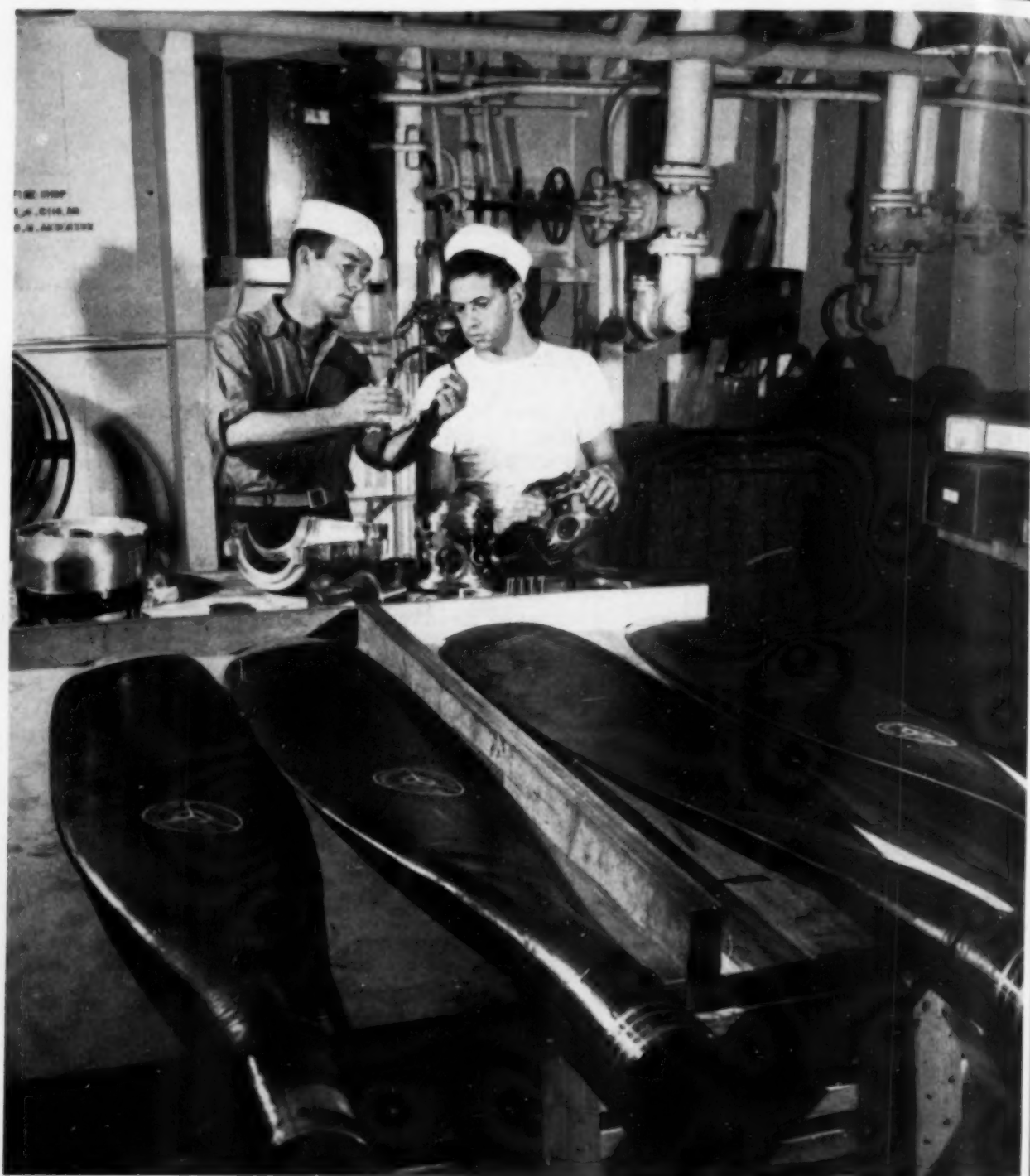
Give us not only the needed implements of war, but the assurance and backing of a united people so necessary to hasten the victory and speed the return of your fighting men.

William B. Dealey
Douglas C. E. King
Dwight D. Eisenhower
C. W. Nimitz
H. H. Arnold

type of trouble more easily than a trouble shooter can.

Trouble Chart

To summarize this entire discussion, the points have been tabulated in a trouble chart, page 1117. In using this it should be emphasized that the "remedies" apply only when all the other factors are under control and are approximately right. One must not, for example, change tools, cutting oil, heat treatment all at once (as is too often done at the instigation of production men) when the mounting is not rigid enough — nor, for example, should the metallurgist insist that hardness or microstructure or both is the one criterion of machinability. 



Propeller Shop Aboard Repair Ship for B-29's

It is now well known that one of America's potent "secret naval weapons" is the group of repair and supply ships which are the equivalent of a naval base that follows the fleet. Similarly the Air Technical Service Command has "Aircraft Repair Units (Floating)" that service the aircraft on islands in the Pacific during the early months

of occupation and until adequate shops can be erected ashore. Such a ship is equipped to do all repairs on even the largest aircraft, other than engine overhauling. Traffic between deck of floating shop and damaged aircraft is by launches, ducks, jeeps, helicopters—even diving bells. Photo by Air Technical Service Command.

MAXIMUM CARBON IN CARBURIZED CASES

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Theoretical considerations indicate that the maximum carbon in a carburized case should be no higher than the A_{cm} point for the carburizing temperature, but very much higher carbons are frequently found in practical operation. The author explains this discrepancy by relating excess carbides in the case to the undiffused complex carbides or carbide nuclei in the steel during the carburization; therefore undesirable surface carbides can be avoided if the steel is first converted to truly homogeneous austenite.

THE QUESTION of maximum carbon contents in carburized cases has received considerable attention with the publication of Schlumpf's¹ and subsequently of Harris's² results. There are two schools of thought on this subject. One, advocated by McQuaid, Boegehold and others, maintains that extremely high carbon contents (2.1 to 3.0% C) may be obtained as a normal phenomenon, and they explain this condition by disparate rates of carbon absorption and carbon diffusion in the steel. The second school, of which Harris is the chief exponent, maintains that under properly controlled furnace conditions the maximum carbon content in the case comes fairly close to the amount called for

by the A_{cm} line of the iron-carbon equilibrium diagram, modified according to alloy content. Furthermore, Harris claims that the excessive carbon build-up is an abnormal condition.

There is considerable amount of experimental evidence to support both views. Schlumpf's results show excessive carbon contents at 1500 and at 1600° F. when carburizing S.A.E. 3115 steel (3.1% C, 1.8% C respectively). McQuaid, in *Metal Progress* for July 1944, p. 97, reports several cases of high carbon content (a) when carburizing S.A.E. 5130 steel at 1750° F. for 8 hr. and then dropping the temperature to approximately 1500° F., and (b) when carburizing Ni-Mo gears at 1700° F., cooling to 1200 and re-carburizing at 1500° F. for about 4 hr. Manning, in *Transactions* for March 1943, p. 8, reports excessive carbon contents for Cr-Mo and Ni-Cr steels.

Harris, on the other hand, in support of his view, cites several carefully conducted experiments in which he first decarburized the surface and then carburized it under a gas atmosphere. The maximum carbon content was close to that called for by the A_{cm} line, modified according to alloy content.

The question now arises "What is the reason for the apparent discrepancy?" It is claimed that in some of Harris' tests, oxygen was probably absorbed during decarburization and might have altered the surface of the steel. On the other hand, lack of control and temperature fluctuations are blamed for the high carbon build-ups. In some instances high carbon is explained away

1. H. W. McQuaid, "A study of the Effect of the Aluminum Addition on the Structure of a Quenched Carbon Steel", *Transactions*, 1937, Vol. 25, p. 490.

2. F. E. Harris, *Metal Progress*, April 1944, p. 683. For a complete list of his writings see *Metal Progress*, April 1945, p. 713.

simply with the statement that certain steels, notably chromium steels, are prone to produce this condition. Thus, considerable confusion exists with respect to the problem of maximum surface carbon content, and a clarification of this issue is essential.

A careful examination of the facts divulged in recent literature shows that excessive high carbon contents are principally obtained under the following two main conditions:

1. When a fluctuating temperature is employed or when the carburizing is done at one temperature and then continued at a lower.

2. When steels containing "stable carbide formers" are carburized.

Now the above two conditions have one thing in common—the tendency for carbides to be present during carburizing. In the first, carbide is rejected from solution as a result of a decrease of carbon solubility in austenite during cooling from the high carburizing temperature to the lower carburizing temperature. (This is on the assumption that the carbon content of the austenite at the high temperature was greater than the saturation value at the lower temperature.) In the second condition, the steel might contain stable carbides or carbide nuclei at the carburizing temperature and retain them during the carburizing operation. This leads directly to the theory which the writer believes explains the question of the excessive high carbon contents:

The maximum surface carbon content in carburized cases depends on the microstructure of the steel during the process of carburization. Excessive carbon contents will be obtained in the case when the steel contains "free" carbides or carbide nuclei during carburization. Carbon contents called for by the A_{cm} line modified according to alloy content will be obtained when the steel is homogeneous austenite during carburization.

The above theory was then checked by several simple tests. Two rectangular pieces of S.A.E. 6152 steel (1% chromium, 0.12% vanadium, pearlitic structure), ground on all surfaces and polished with 0000 metallographic paper, were squeezed tightly together and held thus in

Fig. 2—Microstructure of the Other Side of the Same Specimen as in Fig. 1. This side was in close contact with another specimen of the same steel for 15 hr. before being exposed to the carburizing salt. No coarse carbides are present along this surface; the fine dark grain boundary outlines (carbides) are due to a slack quench

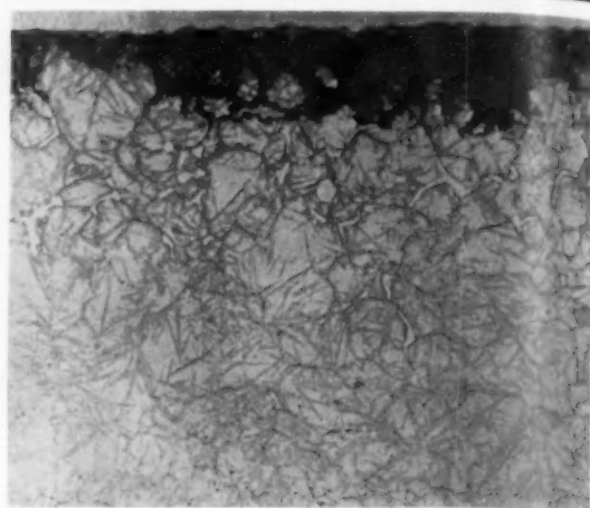
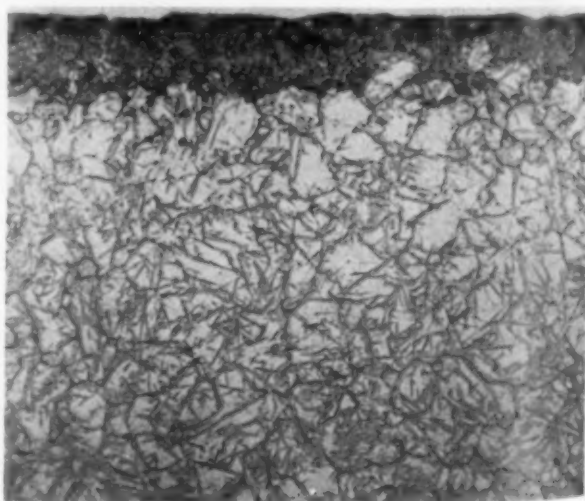


Fig. 1—Coarse Carbides in the Microstructure of One Side of an S.A.E. 6152 Steel Specimen Exposed to a Carburizing Salt Bath (8% CN) at 1700° F. for 8.5 Hr. Etch: 4% picral; \times 500

a liquid carburizing salt bath (8% CN) at 1700° F. for 1½ hr. in such a manner that no salt could penetrate between the two surfaces. At the end of the time the two pieces were separated under the salt and left in the bath for seven additional hours. The two pieces were then quenched simultaneously in oil. Cross-sections were polished and etched for microscopic examination, with results shown in Fig. 1 and 2. The side exposed to the salt for the full time (9½ hr.) shows coarse, "free" carbides, while the other side which, in essence, was given a preliminary austenizing treatment for 1½ hr., does not show this condition.



The above test was then repeated using NE8720 steel. The temperature of carburization was 1525° F. and a salt bath containing 26% CN was used for carburizing. The samples were held together for 1½ hr. before separating and were then left in the bath for 16 hr. additional. Microstructures are shown in Fig. 3 and 4. Again, the side which was austenized prior to carburization did not show any coarse carbides, while the side which was exposed directly to the carburizing action did.

In either of the above tests, the difference in the actual time of carburizing of the two sides cannot account for the difference in microstructure. As proof of this statement, reference is made to Fig. 5, which shows coarse carbides in NE8720 steel when carburized only 3 hr. in a 26% CN salt bath at 1525° F. Furthermore, it cannot be claimed that the surface which received prior austenization could have been altered in any other manner, since a tight fit existed between the two pieces and since the samples were not

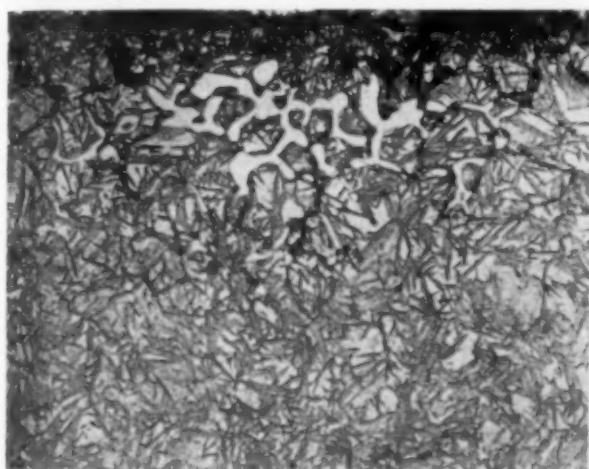
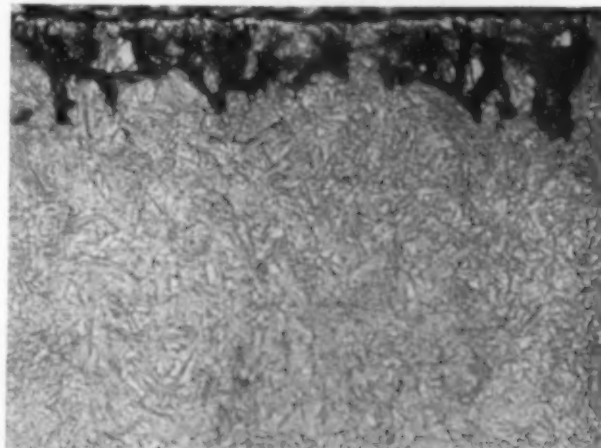


Fig. 3—Coarse Carbides in the Microstructure of One Side of an NE8720 Steel Specimen Exposed to a Carburizing Salt Bath (26% CN) at 1525° F. for 17.5 Hr. Etch: 4% picral; × 500

removed from the salt bath during the entire time. The difference in the microstructure could therefore be due only to the fact that one side was homogeneous austenite when exposed to the carburizing medium. (It is assumed that 1½ hr. is sufficient for full austenization of NE8720 at 1525° F. and of S.A.E. 6152 steel at 1700° F.) It follows that the microstructure at the time of carburizing determines whether coarse "free" carbides exist in the subsequent case. This would explain the high carbon obtained when carburiza-

Fig. 4—Microstructure of the Other Side of the Same Specimen as Fig. 3. This side was in close contact with another specimen of the same steel for 1.5 hr. before being exposed to the carburizing salt, and no coarse carbides are present here



tion is partially conducted at one temperature and then continued and finished at a lower temperature. It would also explain why certain steels, such as chromium steels and chromium-molybdenum steels, are more prone to produce high surface carbon content in the case, for the strong "carbide formers" built up by chromium and molybdenum increase the difficulty of full austenization; the complex carbides dissociate and dissolve slowly. Such steels when carburized at lower temperatures (1500° F.) would in all

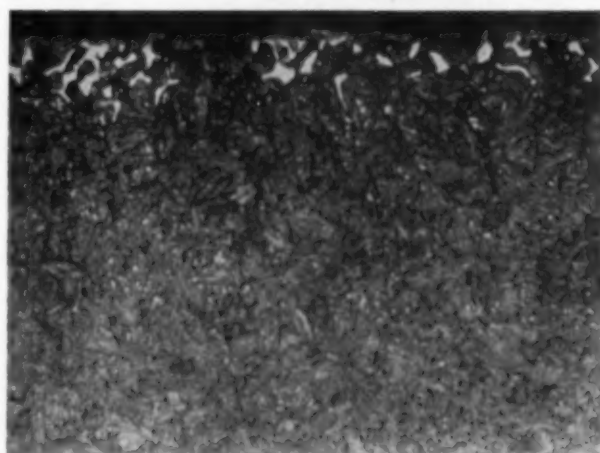


Fig. 5—Coarse Carbides in the Microstructure of an NE8720 Steel Specimen Exposed to a Carburizing Salt Bath (26% CN) at 1525° F. for Only 3 Hr.; Etch: 4% Picral; × 500

probability contain carbide nuclei and would therefore produce surfaces with high carbon contents, while at high temperatures (1700° F.) the rate at which full austenization would be approached might be sufficiently high that no excess carbides will be formed.

As further evidence in support of the proposed theory, the following additional data are submitted:

1. Tests on S.A.E. 6152 steel were repeated, using a 26% CN bath at 1525° F. (The first experiment was in an 8% CN bath at a higher temperature — 1700° F.) The two pieces of steel were held together 1½ hr., were then separated under the bath and carburized for 16 hr. longer.

in this experiment were affected by prior austenization.

2. A plain carbon toolsteel containing 1.02% Mn, 0.20% Cr, and of a spheroidized structure, was carburized for 17½ hr. in a 26% CN bath at 1525° F. and then quenched in oil. The composition of this steel is such as to retain carbide spheroids permanently at 1525° F.; notice, in Fig. 8, the spheroids in the interior of the specimen after the quench, beyond any possible effect from diffusing carbon. Coarse spheroidal carbides are present near the surface. It should be noted that the carbides do *not* occur along grain boundaries, and that in many areas they accumulate in bands. The explanation for these conditions is very likely



Fig. 6—Coarse Carbides Along Grain Boundaries and Within the Grain in the Microstructure of One Side of an S.A.E. 6152 Steel Specimen Exposed to a Carburizing Salt Bath (26% CN) at 1525° F. for 17.5 Hr. Etch: 4% picral; × 400

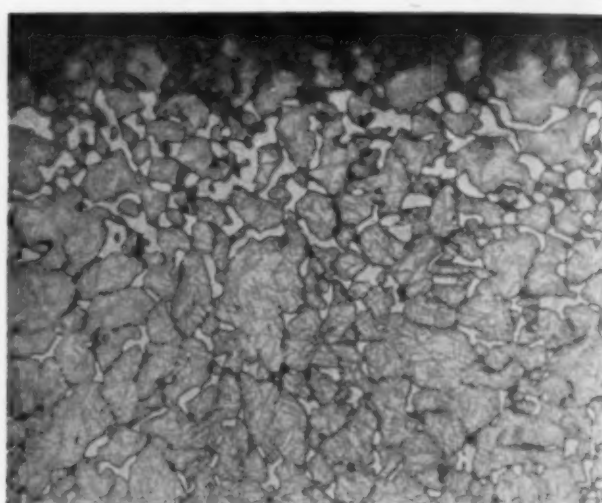


Fig. 7—Coarse Carbides Along Grain Boundaries in the Microstructure of the Other Side of the Same Specimen as Fig. 6. This side was in close contact with another specimen of the same steel for 1.5 hr. before being exposed to the salt

Microstructures are shown in Fig. 6 and 7. Both sides show carbides, but even so the difference is significant. The side that had been austenized shows far less carbides and these occur mainly along grain boundaries. The other side shows more carbides, which near the surface are practically massive, and then slightly deeper they occur within the grain as well as along grain boundaries. Beneath the surface about 0.005 in., in the region which was at temperature for some time before much carbon could diffuse into it, the microstructure is similar to the austenized specimen at the surface. It therefore appears that the amount and location of the carbides formed by diffusion are related to the amount and location of pre-existing carbides and carbide nuclei, which

to be found in the location of the original carbide spheroids, and in the original banded structure of the steel after rolling.

The above two tests, in addition to showing the effect of the presence of carbides on the microstructure after carburizing, tend to bear out the theory in that they relate the amount and location of the carbides formed during carburizing with the microstructure of the steel as it exists during carburization.

We can now look more closely into the probable picture of what is happening during carburizing as a further aid in understanding the reason for the reported discrepancies:

The tendency in any physico-chemical system is to establish equilibrium conditions. As the

steel increases in temperature above the $A_{c1.3}$ point, the carbides will diffuse into the austenite until homogeneity is established, given sufficient time. However, in steels containing alloying elements known to be stable "carbide formers", complete diffusion of the alloying element as well as the carbon must occur before the austenite is fully homogeneous. Since carbide formers diffuse very slowly*, the steels containing such alloying elements may contain regions of either carbide particles or carbide nuclei during the entire carburizing cycle. (These regions will be referred to as "regions of incomplete austenization".) While these regions are in process of austenization another process is in operation at the same time, namely, carbon diffusion inward from the surface. As the austenite becomes saturated with respect to carbon in the vicinity of the regions of incomplete austenization, the conditions are suitable for carbide precipitation and growth. (Any minor fluctuation from the exact equilibrium conditions would precipitate carbide.)

Carbide precipitation and growth during the carburizing period may also be explained on the basis of the equilibrium diagram. As stated above, the region in the vicinity of the carbide particles or nuclei becomes saturated with respect to carbon. The result is that this area now finds itself past the A_{cm} line in the austenite+cementite area of the equilibrium diagram. The conditions are therefore suitable for further carbon absorption until the cementite line of the equilibrium diagram is reached.

The theory proposed by the writer states that "carbon contents called for by the A_{cm} line, modified according to alloy content, will be obtained when the steel is homogeneous austenite during carburization". No definite experimental proof has been offered in this article to support this statement. It is based essentially on the data presented by Harris in his experiments on steels with decarburized surfaces. By decarburiz-

ing the steel prior to the carburizing treatment he merely eliminated the possibility that carbide nuclei could be present during carburization, and he obtained carbon contents called for by the A_{cm} line. In addition it may be argued, on the basis of the tests described herein, that since the formation of coarse carbides depends on the nucleating effect of pre-existing carbides, the absence of such pre-existing carbides (as in fully austenized steel) would preclude the formation of coarse carbides. Consequently, the maximum carbon content in such cases can correspond at most only to the saturation value of the austenite.

Effect of Carburizing Medium—Schlumpf used pack carburizing in his tests, while Harris

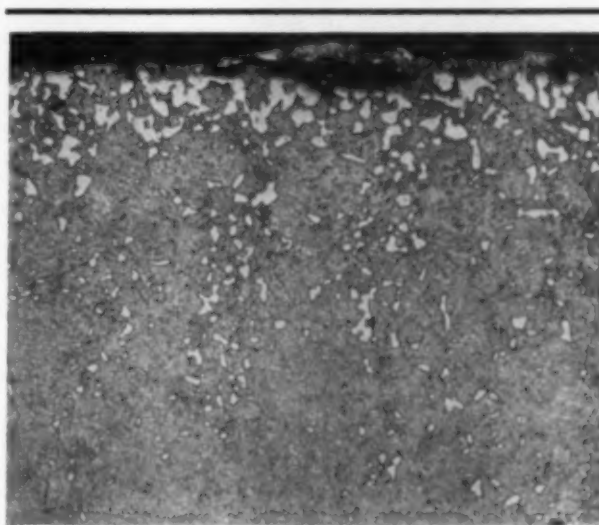


Fig. 8 — Coarse Carbides in the Microstructure of a Spheroidized Plain Carbon Toolsteel Specimen Exposed to a Carburizing Bath (26% CN) at 1525° F. for 17.5 Hr. Etch: 4% picral; $\times 250$

carburized his specimens in a gas atmosphere. The results described in the preceding paragraphs were obtained from a liquid cyanide bath. In order to show that the carburizing medium, as such, is of no importance insofar as the validity of the proposed theory is concerned, the tests using S.A.E. 6152 steel were repeated under a gas atmosphere in a pit type furnace. The two samples were held together by wire and suspended through the blast gate by two drill rods (one rod attached to each sample as shown in Fig. 9). To insure that the surfaces in contact with each other were not being subjected to the carburizing action for the first $1\frac{1}{2}$ hr. at heat (1725° F.), the edges were mud-packed and wrapped with asbestos tape. After $1\frac{1}{2}$ hr. the two samples were separated and carburizing was continued for an

*E. C. Bain in his book on "Functions of the Alloying Elements in Steel" discusses this problem quite fully in his Chapter IV on "Effects of Alloying Elements in Forming Austenite". For example (p. 128): "In a particle of molybdenum-rich carbide, the so-called omega phase, the molybdenum content may be 80 to 90%. Regardless of the constitution of the steel as a whole, molybdenum must diffuse away until only perhaps 8%, at a maximum, remains at such a location before austenite may form there. Such a degree of diffusion may take place in reasonable time only at the highest possible heating temperature.... This matter of degree of solution of slowly dissolving alloying elements is an important one which almost certainly accounts for variable properties secured from the same steel under different conditions."

additional 7 hr. The microstructures on the two sides are shown in Fig. 10 and 11.

The results are exactly similar to those obtained from the salt bath. It is evident, therefore, that the proposed theory holds irrespective of whether gas, liquid, or the usual pack carburizing is used.

Effect of Carburizing Potential—The mechanism outlined in the preceding paragraphs requires the saturation of the austenite with respect to carbon as a condition for carbide precipitation. The correctness of this condition may be seen from the argument that if the austenite in the vicinity of the regions of incomplete austenization were not saturated, these regions would diffuse into the austenite instead of growing into coarse carbides. On this basis, the importance of the "carburizing potential" of the carburizing medium becomes apparent immediately. ("Carburizing potential" is defined as the available carbon for diffusion into the steel, and may be considered as the surface carbon content of homogeneous austenite in equilibrium with the carburizing atmosphere at the carburizing temperature.)

If the carburizing potential of the medium used is sufficient to produce saturated austenite at the carburizing temperature, the excessive carbon build-up will depend on the rate of full homogenization of the austenite. If the carburizing potential is low, no excessive carbon contents will be obtained, irrespective of the rate of austenite homogenization. This explains Manning's results, that "coke containing 15% Ba appeared to overcome completely the tendency of chromium toward a high carbon surface". It should be noted, however, that the carburizing potential of this medium is such as to produce a surface carbon of only 0.85% when carburizing plain carbon steel at 1750° F. (The saturation value of austenite at this temperature is approximately 1.3% carbon.)

This point of carburizing potential was further checked by the writer. Two pieces of NE8720 steel were carburized at 1740° F. for 9 hr. in a salt bath containing 8% cyanide. By an experiment noted two paragraphs further on, it would be expected that these specimens, if quenched directly, would not have contained excess car-

bides at the surface. One piece was then transferred into a 6% cyanide bath at 1550° F., while the other one was placed directly into a 2% cyanide bath held at the same temperature as the 6% bath (1550° F.). After 4 hr. in these secondary baths both samples were quenched. The sample transferred to the 6% bath (re-carburized at lower temperature and in a bath of slightly lower potential) showed coarse "free" carbides, while the one from the 2% bath did not show this condition. The only difference was in the potential of the two baths; the 6% cyanide bath has a much greater carburizing potential than the 2% bath. The difference in the results appears more remarkable when attention is called to the fact that grain boundary carbides precipitated out when the temperature was lowered to 1550° F. but they disappeared during the stay in the 2% bath.

Effect of Temperature—On the basis of the proposed theory, let us now analyze the possible effect of temperature. Assume two temperatures, one T_L low (say 1600° F.), the other T_H high (say 1725° F.). An increase in temperature increases the rate of homogenization. If the rate of austenization is sufficiently rapid at T_L , the lower temperature, the maximum carbon contents at the surface of the cases produced at the two temperatures will correspond to the respective A_{cs} values and consequently a higher carbon case will be expected at T_H , the high temperature. However, if the rate of austenization

is slow at T_L , the low temperature, but rapid at the high temperature T_H , the surface carbon content at T_L , the lower temperature, will actually be higher. (Compare Fig. 1 and 5. This also corresponds to Schlumpf's results quoted by McQuaid: Carburizing at 1500° F. gave a case whose maximum was 3.10% carbon at the surface cut; carburizing at 1600° gave a case with 1.80% max. carbon; 1700° gave a 1.20% max. carbon case.) If the rate of austenization is slow at both temperatures, high carbon contents will be obtained in both cases, but due to the higher diffusion rate at higher temperature, the high carbon contents will exist to a greater depth after carburizing at the high temperature. However, due to the normally higher austenization rates at high temperatures, it is again possible to obtain a greater

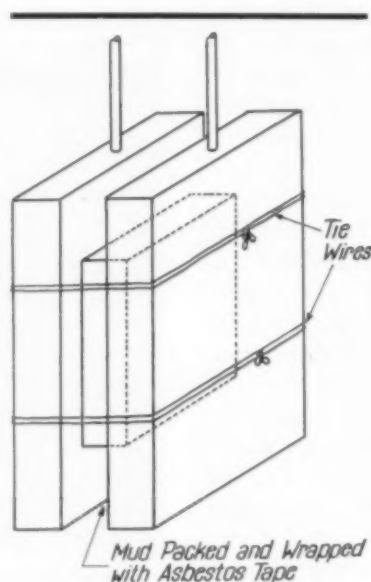


Fig. 9—Arrangement of Experiment in Gas Atmosphere

carbon content near the surface at the lower temperature. This may result from the fact that a relatively higher percentage of the steel will be fully austenized at the higher temperature.

With respect to temperature, mention should be made of some additional tests. NE8720 steel, when carburized at 1700° F. in 8% CN carburizing salt bath, did not show any carbides even when the prior austenization treatment was omitted. This result, in addition to the data on the S.A.E. 6152 steel at the two temperatures, helps to illustrate the effect of temperature discussed in the preceding paragraph.

have furnished direct proof and a clear theoretical explanation of these facts which even slight oscillations in the temperature produce on the apparent limit of saturation of the iron by the carbon in the cementation processes."

From the above, it is apparent that Giolitti offered no explanation for the occurrence of excessively high surface carbon contents at constant temperatures. In the tests conducted by the writer, there can be no question of temperature fluctuations, since the difference in the results was obtained on the same specimen. The theory of fluctuating temperature, therefore, is

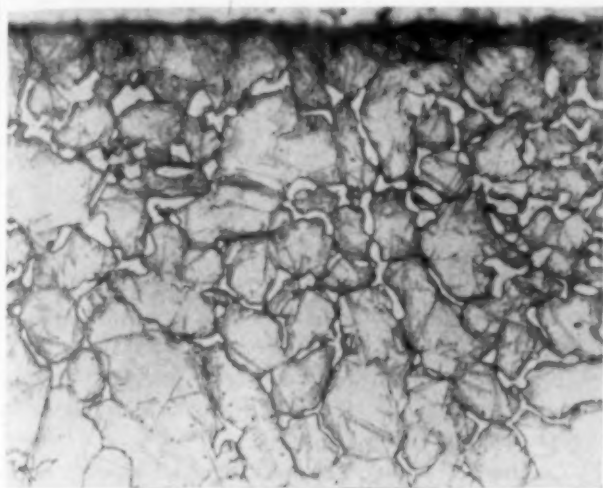


Fig. 10 — Coarse Carbides in the Microstructure of One Side of an S.A.E. 6152 Steel Specimen Exposed to a Carburizing Gas Atmosphere (2.0% CH₄, 20.4% CO, 0.2% O₂, 45.6% H₂, 32.6% N₂, and 10° Dew Point) at 1725° F. Etch: 4% picral; × 500. Compare with Fig. 1, the result of cyanide bath treatment

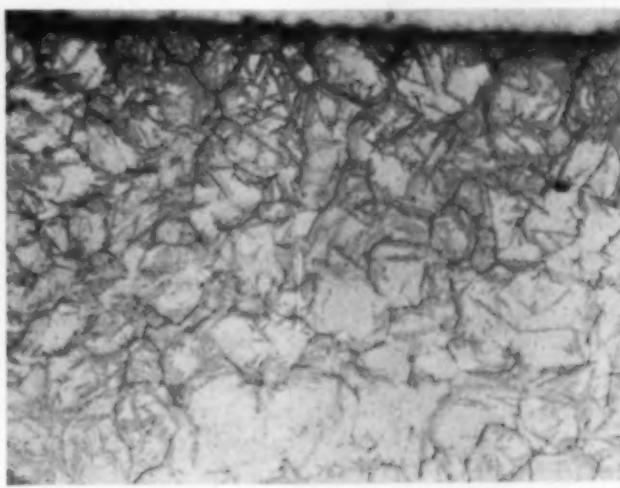


Fig. 11 — No Coarse Carbides Present in the Microstructure of the Other Side of the Same Specimen as Fig. 10. This side was in close contact with another specimen of the same steel for 1½ hr. before being exposed to the gas atmosphere. Compare with Fig. 2

No discussion on the maximum surface carbon content in carburized cases would be complete without mentioning the classic work of Giolitti. In his book "Cementation of Iron & Steel" Giolitti considers this question at length. He maintains that "the occurrence of these phenomena, where under ordinary conditions the formation of free cementite cannot take place, is due to the oscillations of temperature of cementation." To quote further: "Benedicks showed the impossibility of attaining by means of cementation a definite maximum concentration of the carbon corresponding to saturation of the iron, except by keeping the temperature rigorously constant during the whole of the cementation. Still more recent considerations and experiments

not satisfactory to explain the results obtained by the writer. On the other hand, the high surface carbon content resulting from temperature fluctuations is fully accounted for by the theory proposed in this paper.

In conclusion, the writer wishes to point out that if the theory proposed in this paper is definitely established by further tests, it may serve several important functions: (a) Predict by comparison of austenization and carburizing rates whether high carbon contents will be obtained at carburized surfaces; (b) obtain excessive carbon contents at will; (c) obtain any desired size and distribution of carbides in carburized cases; and (d) provide a new method for studying austenization rates.

HEAT TREATMENT OF R301 ALLOY

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SEVERAL ARTICLES recently published have given the general characteristics of the alloy, R301. (See references No. 1 and 2 on page 1132.) The present article, covering its heat treatment in detail, has been prepared in response to numerous requests for additional information on the heat treating practices and their effect on the properties.

R301 is a composite material, constituted of a core of high strength aluminum alloy, clad with a corrosion-resistant aluminum alloy. The composition of core and cladding, as set by the recently published Army-Navy specification AN-A-22, is as follows:

ELEMENT	CORE	CLADDING
Copper	3.9 to 5.0%	0.10% max.
Silicon	0.5 to 1.2	0.35 to 1.0
Manganese	0.4 to 1.2	0.75 max.
Magnesium	0.2 to 0.8	0.8 to 1.5
Iron	1.0 max.	0.6 max.
Chromium	0.25 max.	0.35 max.
Zinc	0.25 max.	0.20 max.
Titanium	—	0.10 max.
Other elements, each	0.05	0.05 max.
Other elements, total	0.15	0.15 max.
Aluminum	Remainder	Remainder

R301 is produced in three tempers, namely, the annealed temper, the solution heat treated temper, and the solution heat treated and aged temper. These three tempers are designated respectively O, W, and T. Typical microstructures of the three tempers are shown in the data sheet, page 1136-B, and are intended to serve as a guide in the micro-examination

of R301 material. (Microstructure of R301-T, not etched, is omitted since it does not differ from that of R301-W, not etched.)

The minimum guaranteed and the average mechanical properties of these three tempers are reported in Table I, page 1130.

Heat Treating Temperature — Figure 1 shows the effect of the heat treating temperature on the mechanical properties of R301, both in the W and T tempers. The drop in mechanical proper-

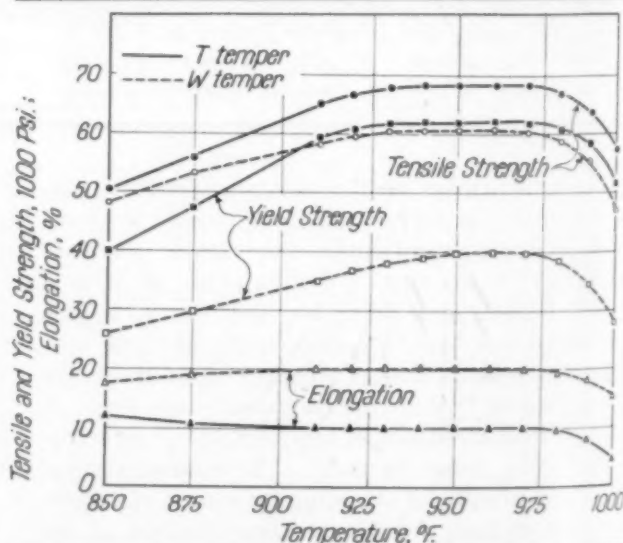


Fig. 1—Effect of Temperature of Solution Treatment on the Tensile Properties of R301-W (Dotted Lines) and R301-T (Full Lines), 0.064-In. Gage

ties caused by eutectic melting in the alloy does not take place until a temperature of 970° F. has been exceeded. Micro-examination has determined that the most fusible constituent in R301 is the Al-CuAl₂-Si eutectic, shown in Fig. 3, whose melting point is 977° F. (Reference 3). Temperatures up to 970° F. can therefore be used safely for heat treatment. However, the mechanical properties do not substantially increase with heat treating temperatures above 940° F. Moreover, during solution treatment the copper in the core material tends to diffuse through the cladding, lowering the corrosion resistance. The higher the heat treating temperature, the higher

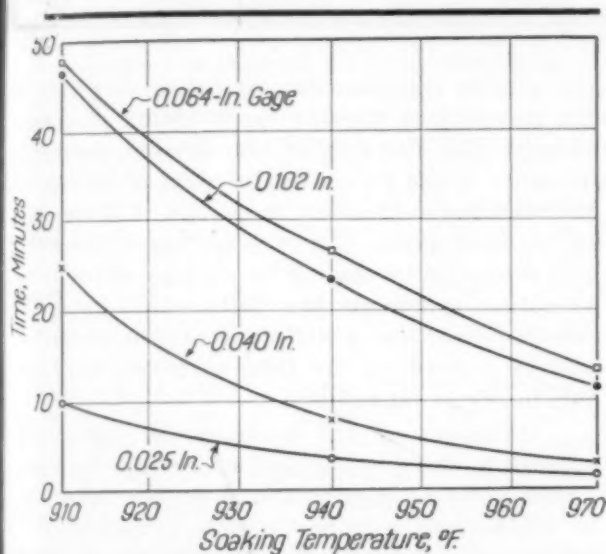
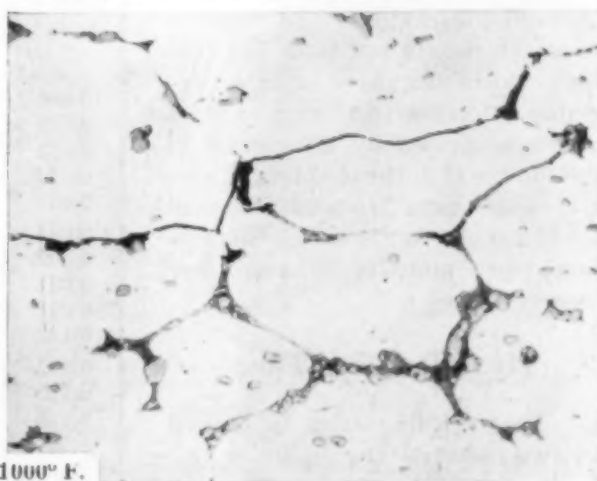
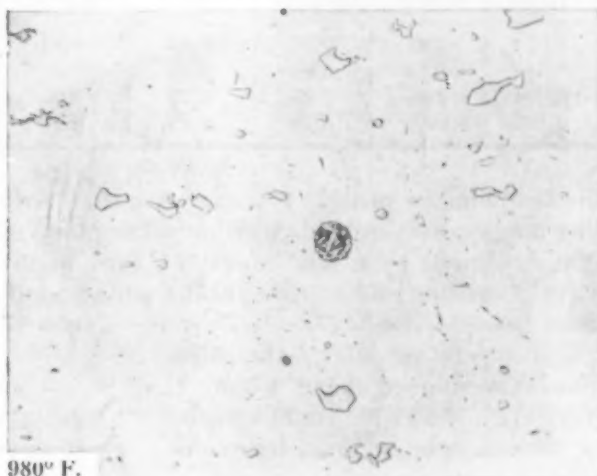
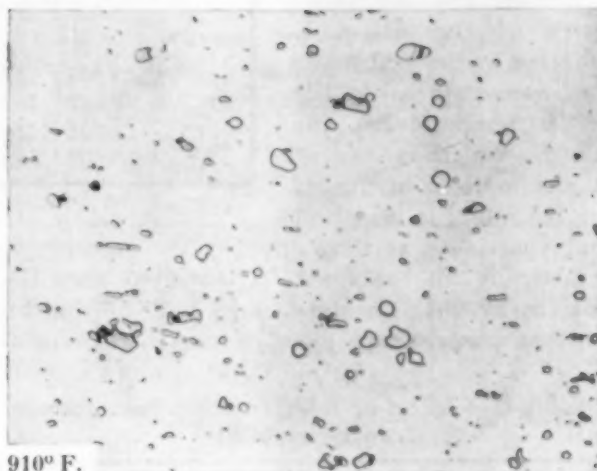


Fig. 2 — Effect of Soaking Temperature on Diffusion — That Is, Minutes Required for Copper to Build Up to 0.05% at the Outer Surface of the Cladding on Various Gages of R301 Sheet

is the rate of diffusion, as shown in Fig. 2. In addition, R301 (as all other materials) when quenched from higher temperatures is susceptible to more buckling. For all these reasons the recommended heat treating temperature has been set at $940 \pm 10^\circ$ F.

The heat treating temperature also has some effect on the formability of the alloy, as indicated by the Erichsen test. Table II gives Erichsen values for some gages of R301-W heat treated at

Fig. 3 — Unetched Structures ($\times 500$) of Core of R301 After Various Solution Heat Treatments. Some CuAl₂ is undissolved at 910° F.; a rosette of Al-CuAl₂-Si eutectic appears at 980° F., and eutectic melting at grain boundaries at 1000° F.



different temperatures. Although there is not a wide difference, and some scattering of the results tends to mask the effect, it can be seen that lower temperatures are conducive to somewhat better formability.

Naturally, the structure of the alloy is affected by the heat treating temperature. At the lower temperatures, the CuAl_2 , which is one of the soluble constituents of the alloy, is dissolved only partially, as shown in Fig. 3. At 940° F. practically all the CuAl_2 is dissolved (see the micros in the data sheet, page 1136-B) and no further changes take place in the structure until

Table II — Effect of Heat Treating Temperature on Formability

GAGE	850° F.	875° F.	940° F.	970° F.
0.025	7.87	7.58	7.63	7.52
0.040	8.23	8.06	7.88	7.86
0.064	8.09	7.97	8.13	8.02
0.102	5.89	5.70	4.52	4.68

the temperature of 977° F. is surpassed. Above this temperature, eutectic melting takes place, at first evidenced by a few "rosettes", then in the form of melting and cracking at the grain boundaries (Fig. 3, page 1129).

Temperature affects the diffusion, as mentioned above. Figure 2, page 1129, shows the time required at various solution heat treatment temperatures for copper to diffuse through the cladding and reach a concentration of 0.05% at the surface. Although there is an appreciable difference from gage to gage, which is caused by differences in the thickness of the cladding (Table III), the higher temperatures will cause copper to appear at the surface more quickly than at lower temperatures.

Heat Treating Time

Heat treating time is limited by two factors: On the lower side is the necessity of dissolving all

Table I — Mechanical Properties of R301

TEMPER	GAGE	MINIMUM GUARANTEED			AVERAGE		
		YIELD, PSI.	TENSILE, PSI.	ELONG., % IN 2 IN.	YIELD, PSI.	TENSILE, PSI.	ELONG., % IN 2 IN.
O	All	—	<30,000	16.0	10,000	25,000	22.0
W	<0.040	37,000	56,000	14.0	39,000	59,000	18.0
	>0.039	37,000	57,000	15.0	41,000	60,000	19.0
T	<0.040	56,000	63,000	7.0	59,000	66,000	9.0
	>0.039	57,000	64,000	8.0	61,000	68,000	9.0
Re-Heat Treated Material							
W	<0.040	35,000	55,000	15.0	36,000	59,000	20.0
	>0.039	35,000	55,000	15.0	37,000	60,000	21.0
T	<0.040	54,000	62,000	8.0	59,000	66,000	9.0
	>0.039	54,000	62,000	8.0	61,000	68,000	9.0

the soluble constituents; on the higher side is the necessity of limiting the diffusion of copper through the cladding to the lowest possible amount. Table IV shows the effect of increasing times on the mechanical properties of three commonly used gages. An examination of this table will show that increasing the soaking times above a certain minimum has little effect upon the tensile properties; a slight increase in properties can be noticed as the time increases but it is practically of no value.

Diffusion, on the contrary, is appreciably affected by the soaking time, and within the range considered, the percentage of copper appearing at the surface is approximately proportional to the time, as shown by the graph in Fig. 4. (Figure 5 presents three micros showing the progress of diffusion in 0.025-gage material.) Diffusion of copper to the surface has an appreciable effect on corrosion resistance, especially in the light gages, as shown by the results in Table IV.

Table III — Thickness of Cladding of R301

GAGE OF SHEET	THICKNESS ON EACH FACE	
	NOMINAL (IN %)	AVERAGE (IN 0.0001 IN.)
0.020	10	20.0
0.025	7.5	18.7
0.032	7.5	24.0
0.040	5	20.0
0.051	5	25.5
0.064	5	32.0
0.072	5	36.0
0.081	5	40.5
0.091	5	45.5
0.102	2.5	25.5
0.125	2.5	31.2
0.150	2.5	37.5
0.170	2.5	42.5
0.200	2.5	50.0
0.250	2.5	62.5

Although there is some scattering of results in Table IV, as is usual in accelerated corrosion tests, it is evident that the increase in soaking time has a deleterious effect on 0.025-gage material, but much less in heavier gages where the total thickness of the cladding is greater. Accordingly, the heat treatment times for the various gages have been set as listed at the top of the next column.

Proper Heat Treatment Times

GAGE	TIME AT TEMPERATURE	ALLOWABLE RE-HEAT TREATMENTS
0.032 and less	10 min.	none
0.033 to 0.063	15	1
0.064 to 0.101	20	2
0.102 to 0.125	30	2
0.125 to 0.250	30	3
0.250 and heavier	60	3

Quenching Medium—As for all the other wrought aluminum alloys, the quenching medium recommended for R301 is cold water. However,

already mentioned, in normal production it should be possible to use a much faster air quench, and the properties will be higher than those reported in Table V.

The microstructure is appreciably affected by the quenching medium, especially in the W temper. No precipitate is visible at 500 diameters magnification at the grain boundaries of material quenched in cold water, polished and etched with Keller's reagent, whereas in the material quenched in boiling water some precipitate is present at the grain boundaries in the form of tiny black particles. In the air quenched material the pre-

Table IV — Effect of Increasing Soaking Time on the Tensile Properties and Corrosion Resistance of R301-T

SOAKING TIME, MIN.	NOT CORRODED			CORRODED*			LOSS BY CORROSION		
	YIELD	TENSILE	ELONGATION	YIELD	TENSILE	ELONGATION	YIELD	TENSILE	ELONGATION
Sheet of 0.025-In. Gage									
10	59,500	67,800	8.5	59,400	67,500	8.6	0.2%	0.4%	—
25	60,700	68,400	8.5	60,500	68,500	8.5	0.3	—	—
40	60,500	68,700	8.8	59,200	67,600	7.5	2.0	1.6	15.0%
65	60,900	69,100	9.1	57,300	66,500	5.2	6.0	3.8	43.0
80	61,400	69,400	9.3	49,300	53,400	3.1	19.7	24.3	67.0
Sheet of 0.072-In. Gage									
20	60,400	68,200	9.5	60,300	67,900	9.6	0.2%	0.4%	—
45	62,300	68,900	9.6	62,400	68,800	9.5	—	1.6	1.0%
70	63,200	69,100	9.6	62,800	69,200	9.6	0.6	—	—
95	62,800	69,400	9.7	62,600	69,300	9.5	0.4	0.2	2.1
120	65,100	69,800	9.4	64,200	69,000	9.3	1.4	1.1	1.0
Sheet of 0.102-In. Gage									
25	60,700	70,000	10.8	60,800	70,100	10.9	—	—	—
55	61,800	70,800	10.6	61,700	70,700	10.8	0.2%	0.1%	—
85	62,000	70,400	10.9	62,100	70,400	10.7	—	—	1.8%
115	61,900	70,600	10.8	61,600	70,300	10.4	0.5	—	2.9

*Immersed 6 hr. in solution of salt (NaCl) and hydrogen peroxide (H₂O₂)

boiling water, oil or air can be used with no noticeable change in the results.

Table V shows the effect of various quenching media on the mechanical properties, formability, electrolytic potential and corrosion resistance of various gages of R301, both in the W and T tempers. The air quench used in these experiments was produced by blowing air from a small fan on the specimens. The time required to cool the specimens from quenching to room temperature was of the order of 7 to 10 min. This very slow "quench" was used in order to investigate the slowest cooling conditions to be encountered in industrial quenching practice.

As can be seen, quenching in boiling water produces only negligible differences in the properties. However, the slow air quench we used produces appreciably lower mechanical properties than either cold or boiling water. However, as

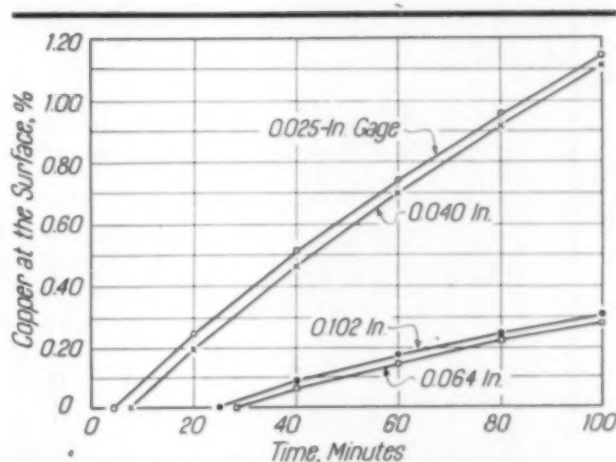


Fig. 4 — Effect of Soaking Time at 940° F. on Diffusion of Copper in Cladding on Various Gages of R301 Sheet

Table V — Effect of Quenching Media on Properties of R 301 in W and T Temper

GAGE (IN.)	QUENCH	NOT CORRODED					POTENTIAL DIFFERENCE, CORE TO CLADDING	CORRODED 24 Hr. IN NaCl + H ₂ O ₂				LOSS BY CORROSION		
		YIELD STRENGTH	TENSILE STRENGTH	ELON. IN 2 IN.	ERICHSEN DUCTILITY	ROCKWELL HARDNESS		YIELD STRENGTH	TENSILE STRENGTH	ELON. IN 2 IN.	YIELD STRENGTH	TENSILE STRENGTH	ELON. IN 2 IN.	
0.025	Cold water Boiling water Air	36,000	58,100	20.2	7.80	E-93.0	159 mv.	36,000	57,400	15.2	0	1.2%	24.6%	
		33,500	57,900	21.5	—	—	—	33,000	56,700	18.0	1.5%	2.1	14.6	
		32,400	56,900	19.2	7.12	E-89.0	120	32,700	55,900	15.4	0	1.7	19.8	
0.040	Cold water Boiling water Air	38,400	61,900	20.4	7.67	E-95.0	164 mv.	38,200	61,500	19.1	0.5%	0.65%	6.4%	
		39,100	62,400	21.0	7.44	E-95.0	129	38,800	61,700	18.3	0.8	1.1	6.4	
		33,700	58,300	18.6	6.76	E-92.5	100	33,100	57,100	14.8	1.8	2.0	20.0	
0.064	Cold water Boiling water Air	39,200	62,200	19.8	8.33	E-97.0	178 mv.	39,100	61,800	18.6	0.2%	0.6%	6.1%	
		38,100	61,600	20.5	8.18	E-96.5	144	37,000	61,000	19.9	2.9	0.8	2.9	
		30,800	55,600	18.0	6.12	E-92.0	100	30,200	54,200	14.5	1.9	2.5	19.4	
0.102	Cold water Boiling water Air	38,900	62,800	21.3	5.22	E-96.0	185 mv.	38,300	62,200	21.0	1.5%	0.9%	1.4%	
		37,900	61,800	19.3	3.78	E-94.5	143	37,500	61,100	18.6	1.0	1.1	3.1	
		27,900	52,400	16.6	3.32	E-89.5	94	27,700	51,300	14.8	0.7	2.0	10.8	
0.025	Cold water Boiling water Air	57,100	64,000	8.9	6.02	E-102	57 mv.	57,000	63,200	7.8	0.17%	1.3%	12.5%	
		57,500	63,600	9.2	—	—	—	57,000	62,400	7.2	0.9	1.9	21.8	
		52,900	61,000	8.8	5.13	E-99	56	51,100	59,000	7.6	3.5	3.3	13.7	
0.040	Cold water Boiling water Air	63,000	70,100	9.7	5.30	E-103.5	49 mv.	63,000	69,600	9.0	0	0.7%	7.2%	
		61,400	68,400	9.1	4.94	E-103.5	50	60,600	67,500	6.7	1.3%	1.3	26.5	
		52,600	62,000	8.6	4.36	E-100	62	52,700	62,300	7.4	0	0	14.0	
0.064	Cold water Boiling water Air	62,500	68,200	10.2	4.86	E-105	67 mv.	61,200	68,000	9.1	2.0%	0.3%	10.8%	
		63,200	69,100	9.8	4.17	E-104.5	79	60,800	67,000	8.4	3.8	2.2	14.2	
		50,200	58,800	9.8	4.17	E-100.5	74	48,400	58,300	8.3	3.5	0.8	15.2	
0.102	Cold water Boiling water Air	63,600	69,500	10.2	2.64	E-104.5	83 mv.	62,200	69,000	9.9	2.2%	0.7%	3.0%	
		62,400	68,600	9.6	2.29	E-103	85	60,800	67,500	8.7	2.6	1.6	9.4	
		46,700	57,900	8.8	2.50	E-98.5	82	45,500	57,200	9.2	2.6	1.2	0	

(a) Heat treated in the laboratory and not flattened thereafter. This accounts for the lower properties, which correspond to re-heat treated material.

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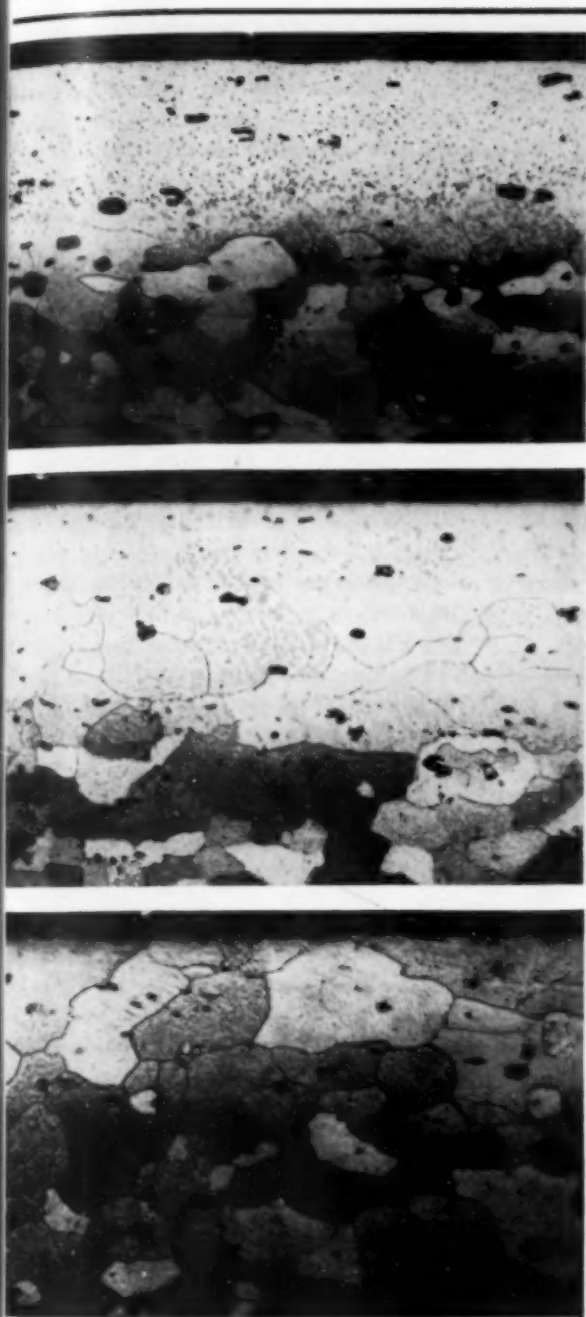


Fig. 5 — R301 (0.025-Gage) After Soaking 10 Min., 40 Min., and 80 Min. at 940° F. (Top, Middle and Bottom, Respectively). Keller's etch: $\times 500$. Soaking 10 min. gives the normal diffusion; the grain boundaries shown in the other two micros, after diffusion has well progressed, exist in the cladding, but the etchant reveals them only when appreciable copper is in solution

cipitate present at the grain boundaries is in the form of platelets.

Formability, as measured by the Erichsen test, is somewhat reduced by the slower quench. However, in actual production, only parts already formed will be slow quenched, this being done to minimize distortion. For this reason the decrease in formability shown in Table V is of little more than academic interest.

Corrosion Resistance Unimpaired

There is some scattering of the accelerated corrosion results, as usual, but it can be seen that the resistance of the alloy is very little affected by the quenching medium, which is in contrast with the results found in other alloys of a similar type (References 4 and 5). This is also confirmed by the electrolytic potential measurements. In the W temper the potential of the core is somewhat affected by the quenching medium, and with the slower quenches the potential difference between core and cladding is somewhat reduced. However, the cladding still gives sufficient electrolytic protection to the core even after the slowest quench. In the T temper no appreciable effect of the quenching medium can be noticed.

This immunity of corrosion resistance to different quenching speeds is very important to the user of R301 for formed parts. Parts which have to be formed in the O temper and subsequently heat treated, tend to distort during quenching in cold water. The slower quenches will cause less distortion and reduce the necessity of restriking or hand straightening. Since the corrosion resistance and the mechanical properties are not decidedly affected by the quenching medium, it will be possible to use slower quenches and obtain parts of correct shape at a lower cost which have approximately the same properties.

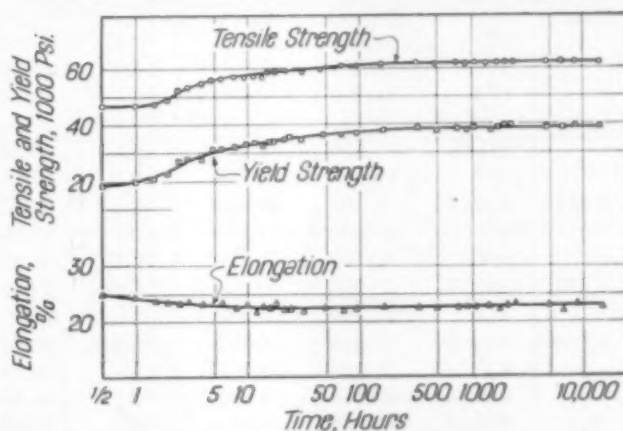


Fig. 6 — Natural or Room Temperature Aging of R301-W, 0.064-In. Gage

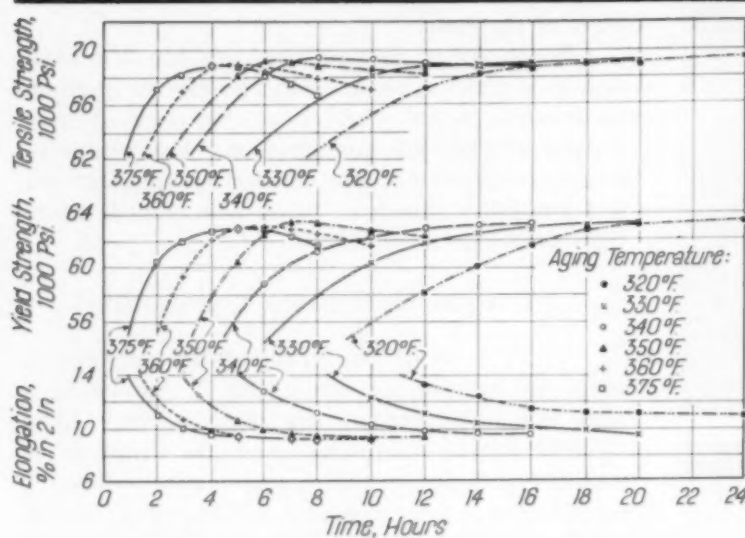


Fig. 7 — Effects of Aging Temperatures and Times on Tensile Properties of 0.064-in. Gage Sheet

Natural or Room Temperature Aging — Figure 6 shows the effect of aging 0.064-in. sheet at room temperature on the mechanical properties of R301 over a period of more than 1 year. As can be seen, the aging starts within 1 hr. after quenching. Most of the aging takes place within the first 24 hr., and after about 1 to 2 weeks the properties of the alloy are completely stabilized; no further change takes place.

Small differences exist in the aging of different gages; as a rule the lighter gages age a little more slowly than the heavier ones, but the difference is not important. For instance, 0.025-gage

material requires usually 2 weeks to become stabilized, whereas 0.102-gage reaches the maximum within 4 days to one week. The heat treatment temperature has also some effect on the aging time; lower temperatures tend to produce faster aging at room temperature than higher temperatures. In any case the differences are very small and are of no practical importance. Table VI shows the mechanical properties of various gages after 4 days and 2 weeks, when heat treated at the different temperatures indicated.

Artificial or Elevated Temperature Aging — A wide range of temperature can be used for artificial aging. The aging time varies with temperature; the higher the temperature, the shorter the time.

Figure 7 shows the effect of time and temperature. Recommended times for these temperatures are as follows:

TEMPERATURE, °F.	TIME
320 ± 5	18 hr. ± 1 hr.
330 ± 5	12 hr. ± 1 hr.
340 ± 5	8 hr. ± 1 hr.
350 ± 5	6 hr. ± 30 min.
360 ± 5	5 hr. ± 30 min.
375 ± 5	4 hr. ± 15 min.

Mechanical properties obtained by the various accelerated aging treatments do not differ to any great extent, so that any of the above recommended practices may be used. Corrosion resistance is also very little affected by the aging treatment. Table VII shows the effect of two

Table VI — Effect of Gage and Heat Treatment Temperature on Natural Aging of R301

GAGE	TEMPERATURE	AGED 4 DAYS			AGED 2 WEEKS			INCREASE IN %		
		YIELD POINT	TENSILE STRENGTH	ELON. IN 2 IN.	YIELD POINT	TENSILE STRENGTH	ELON. IN 2 IN.	YIELD POINT	TENSILE STRENGTH	ELON. IN 2 IN.
0.025	850° F.	28,400	49,500	18.6%	28,900	50,200	18.9%	1.7	1.3	1.6
	875	28,400	50,600	18.8	29,000	51,100	19.1	2.1	1.1	1.5
	940	35,300	58,700	19.9	36,600	59,000	20.0	3.4	0.5	0.5
	970	38,000	60,300	19.7	39,200	61,600	20.7	3.1	2.2	4.9
0.040	850° F.	27,600	48,200	19.0%	27,800	49,500	18.6%	0.7	2.6	—
	875	29,800	51,500	19.8	30,300	52,300	18.9	1.6	1.5	—
	940	38,400	61,700	20.9	38,400	62,200	21.2	—	0.8	1.4
0.064	850° F.	28,000	48,200	19.4%	27,900	49,800	19.2%	—	0.8	—
	875	31,400	51,600	19.7	30,700	52,500	19.7	—	1.7	—
	940	39,100	61,200	20.8	38,600	62,000	21.2	—	1.3	1.9
0.072	940° F.	37,100	60,800	21.0%	37,500	61,100	20.1%	1.1	0.5	—
	970	37,400	60,400	21.0	37,500	61,100	20.5	0.2	0.8	—
0.102	850° F.	28,000	50,800	20.5%	28,600	50,600	20.3%	2.1	—	—
	875	31,400	52,800	19.9	30,300	53,000	19.2	—	0.4	—
	940	39,100	61,700	20.5	38,800	62,000	20.9	—	0.5	2.0
	970	39,800	63,000	20.2	40,000	63,000	20.8	0.5	0.6	3.0

different aging treatments on accelerated corrosion resistance of three different gages of R301-T. Differences in corrosion resistance between the recommended aging treatments (18 hr. at 320, and 6 hr. at 350° F.) are very small, and within the error of testing.

Overaging—If the aging treatment is protracted for a longer time than recommended above, the material tends to be "overaged". However, the time limits are not very critical, as shown by the tolerances noted, within which overaging has no effect on the mechanical properties or the corrosion resistance. Even aging periods far in excess of those recommended have little effect. The third line in each group in Table VII shows the effect of overaging 33% at 320° F. on the mechanical properties and corrosion resistance of the three gages. (Compare the third line in each group with the first line.) Overaging also has little effect on the microstructure, as can be seen by comparing Fig. 8 and the lower right micro in the data sheet, page 1136-B. The precipitated particles at the grain boundaries tend to agglomerate into films, and the precipitate inside the grains begins to appear along the slip planes. The grain contrast (etched) is also reduced.

Delayed Aging—If a long time intervenes between the solution treatment and an aging treatment at temperatures above atmospheric—that is, if a naturally aged material is artificially aged at some distant time, the eventual mechanical properties tend to be somewhat lowered.

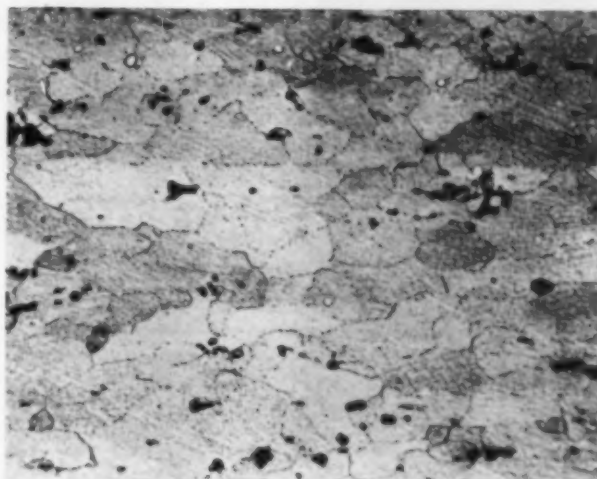


Fig. 8 — Microstructure of R301-T, 100% Overaged, 12 Hr. at 350° F. (Compare with micros in data sheet.) Keller's etchant; $\times 500$

This is shown in Table VIII. This behavior is not peculiar to R301, but is found in all other aluminum alloys. The reasons for it have not been explained completely as yet, although attempts have been made by the authors quoted in references No. 6 and 7.

As can be seen, the effect of delayed aging is more pronounced when the operation is conducted at the higher temperature. Although even a delay of 20 months between quenching and

Table VII—Effect of Aging and Overaging on Corrosion Resistance to NaCl-H₂O₂ Solution

GAGE	AGING		NOT CORRODED			CORRODED 6 Hr.			LOSS IN %		
	TEMP.	HOURS	YIELD	TENSILE	ELON.	YIELD	TENSILE	ELON.	YIELD	TENSILE	ELON.
0.025	320	18	59,300	68,300	9.7	59,300	67,300	9.8	—	1.4	—
	350	6	58,500	67,300	8.5	60,000	67,600	8.3	—	—	2.4
	320	24	59,200	67,800	9.7	59,200	66,400	9.5	—	2.1	2.2
0.072	320	18	63,600	69,500	11.5	64,000	69,000	11.5	—	0.7	—
	350	6	60,400	68,200	9.5	60,600	68,200	9.3	—	—	2.1
	320	24	60,700	68,400	9.8	60,700	67,800	10.0	—	1.0	—
0.102	320	18	60,700	71,000	11.8	62,400	70,100	11.8	—	0.1	—
	350	6	64,800	70,400	10.0	64,000	69,500	10.2	1.2	1.3	—
	320	24	62,000	70,200	10.8	61,700	69,800	10.6	0.5	0.5	1.8

Table VIII—Effect of Delayed Aging

GAGE	AGING		48 Hr. DELAY			10 MONTHS' DELAY			20 MONTHS' DELAY		
	TEMP.	HOURS	YIELD	TENSILE	ELON.	YIELD	TENSILE	ELON.	YIELD	TENSILE	ELON.
0.040	360° F.	5	60,700	68,100	9.0	58,100	66,200	9.5			
	320	18	59,200	69,100	11.7	58,900	67,700	11.7			
0.064	360° F.	5	61,300	68,400	9.7	58,200	66,000	10.8	58,000	65,800	10.7
	320	18	60,300	67,900	11.5	58,900	67,200	12.0	59,000	67,400	11.3

aging does not reduce the mechanical properties below specifications, it is recommended that material in the W temper be stored not over 3 months in order to obtain maximum strength.

Cold Work Prior to Aging—In Fig. 9 and 10 are plotted the properties against the reduction by cold working. They show the effect of cold work on the tensile properties of R301 before and after aging. The reductions plotted were obtained both by cold rolling and by stretching;

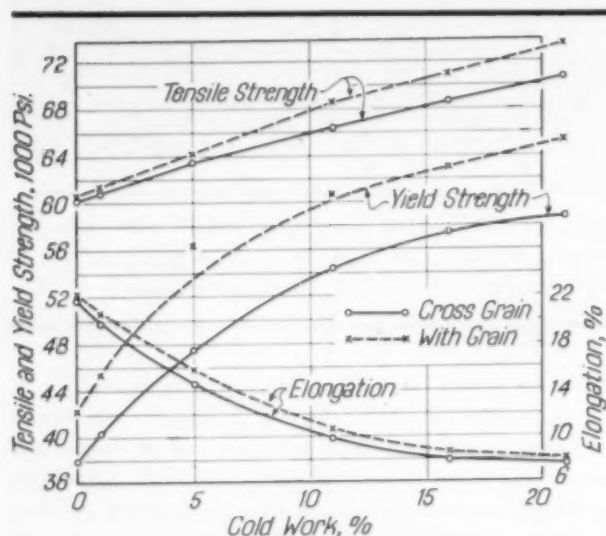


Fig. 9 — Effect of Cold Work in Various Degrees on the Tensile Properties of 0.064-In. R301-W (Solution Treated, not Aged)

no difference was noticed on the effect of the two methods — however, stretching proved to be more difficult and could be used only for the lighter reductions. As can be seen in Fig. 9, the tensile properties of R301-W are appreciably affected by cold work and, with high reductions, closely approach the properties of the material in the T temper.

Figure 10 shows the effect of increasing amounts of cold work prior to aging on the mechanical properties of R301. Small amounts of cold work in this stage (after solution treatment but prior to artificial aging) do not change the properties appreciably, as in the case of other aluminum alloys (References 8 and 9). Even cold working 10% or more does not change the tensile properties to a large extent after the aging treatment recommended.

However, it is well known (Reference 10) that cold working prior to aging tends to accelerate aging. It is to be expected that the material, cold worked to a large extent, requires shorter

aging treatments to develop the maximum properties. Investigations are in progress to determine the facts in this respect.

Heat Treatment of Annealed Material and Re-Heat Treatment—When sheet is heat treated by the producer, a cold water quench is used. This causes appreciable buckling, especially in thin sheet, so that before the material is shipped, it must be flattened and straightened by stretching and rolling, thus cold working the material slightly after solution treatment. The amount of cold work varies somewhat, but as an average it amounts to a reduction of about 1%. This has some effect on the tensile properties in the W temper, as can be seen in Fig. 9. No difference exists in the properties of material in the T temper, since the properties of R301 are not appreciably affected by a small amount of cold work before aging.

When the material is formed before heat treatment, or is re-heat treated, it is not as a rule cold worked after quenching, so that the properties in the W temper will be somewhat lower, as

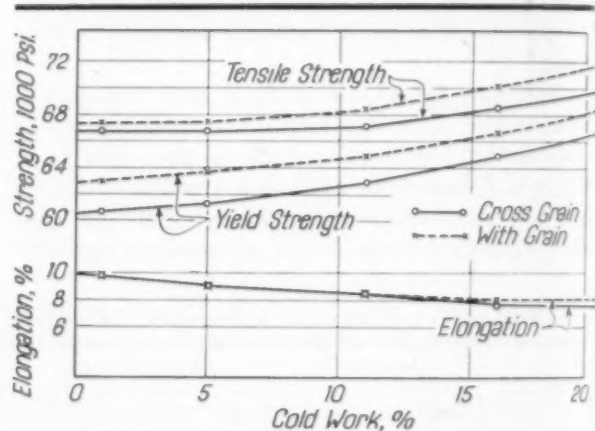


Fig. 10 — Effect of Prior Cold Work in Various Degrees on the Tensile Properties of 0.064-In. R301-T After Aging 6 Hr. at 350° F.

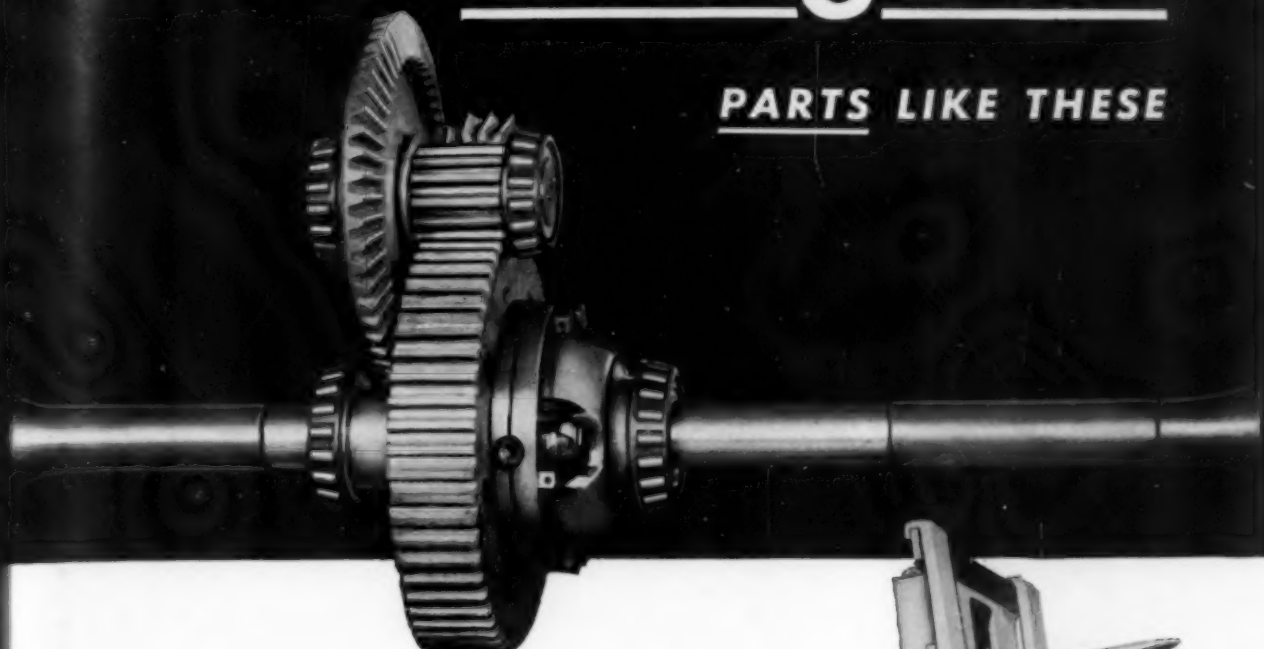
shown by Table I. However, when the same material is aged, the mechanical properties will not differ from those of material purchased in the T temper, although the specification sets a lower limit also for this material.

Acknowledgments—The authors wish to thank the personnel of the Reynolds Alloys and Reynolds Metals companies who have collaborated; special acknowledgment is due to M. A. J. Phillips, W. W. Hairston, L. E. Householder, V. E. Flaherty, H. B. Burrack, M. F. Rupp and R. S. Mapes.

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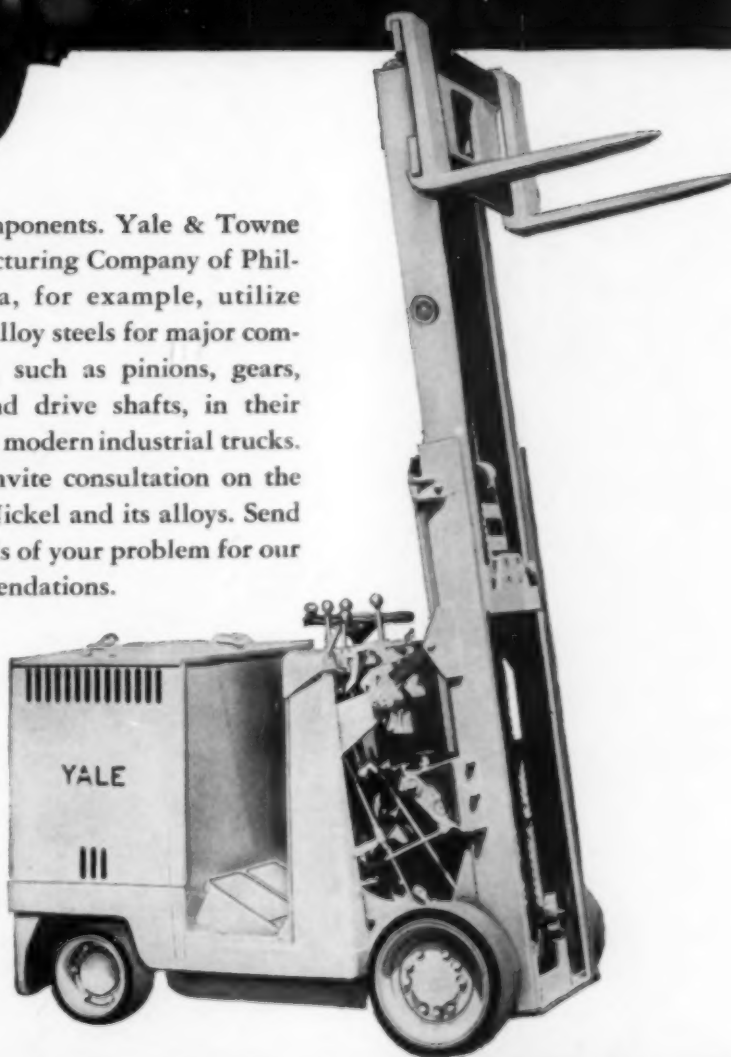
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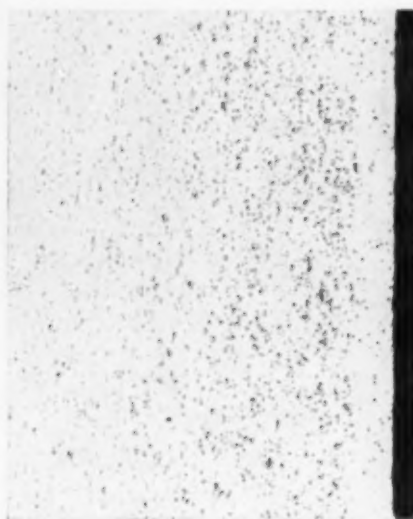


THE INTERNATIONAL NICKEL COMPANY, INC. 67 WALL STREET
NEW YORK 5, N.Y.

Microstructure of R 301

By T. L. Fritzlen and L. F. Mondolfo

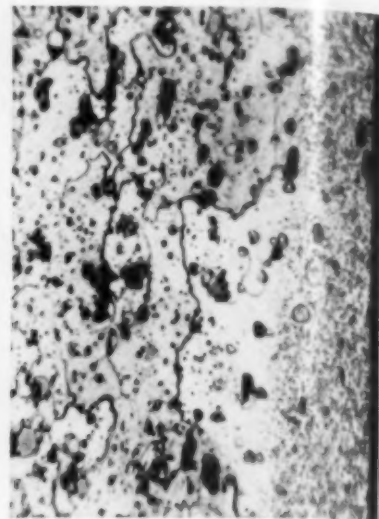
Code: O = Annealed; W = Solution heat treated at 940° F.; T = Solution heat treated and aged.



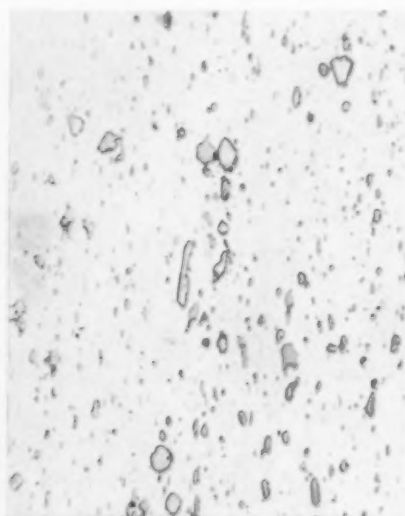
R301-O; not etched. Large amount of undissolved constituents (mainly CuAl₂) in the core material. × 75



R301-O; Keller's etchant. Almost complete lack of grain contrast. × 75
In these views, cladding is at right.



R301-O; Keller's etchant. Grain contrast absent; soluble constituents precipitated in core and cladding. × 500



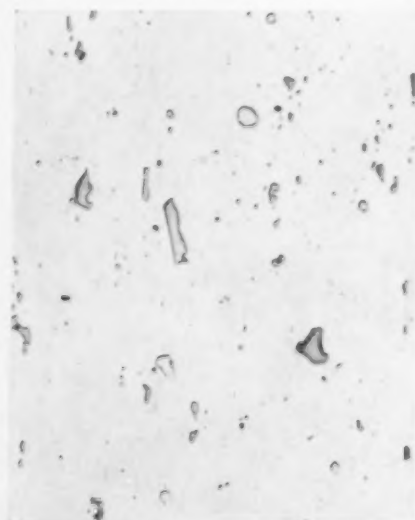
R301-O; (core); not etched. Numerous small particles of CuAl₂ undissolved; several larger, darker particles of Mn constituents. × 500



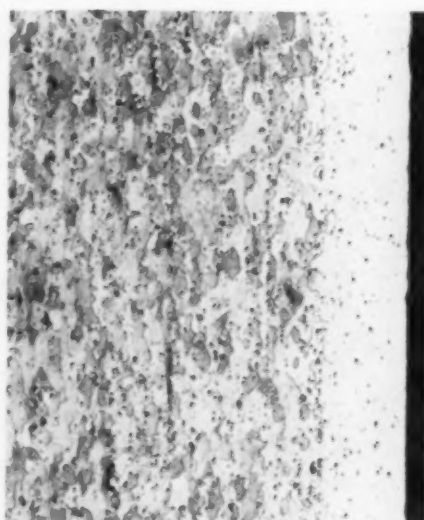
R301-W; Keller's etchant. High contrast between grains. × 75



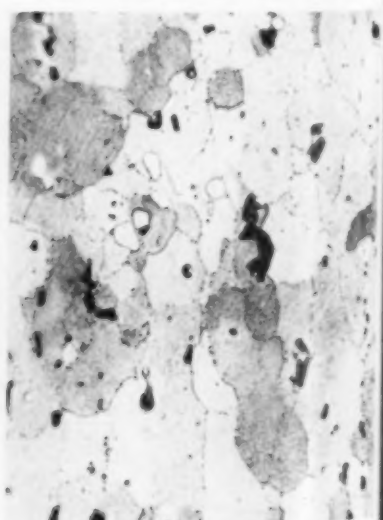
R301-W; Keller's etchant. High grain contrast; no precipitate observable at grain boundaries. × 500



R301-T (or R301-W) not etched. Mn constituents darker; a few particles of CuAl₂ (lighter) still visible. × 500



R301-T; Keller's etchant. Lower grain contrast than in W temper, shown above. × 75



R301-T; Keller's etchant. Low grain contrast; tiny precipitates seen at grain boundaries. × 500

RECLAMATION OF MAGNESIUM CASTINGS BY HELIUM ARC WELDING

By B. L. Averbach
Chief Metallurgist
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SALVAGE of magnesium castings rapidly becomes an economic necessity for a foundry, and fortunately there is a process which allows the foundryman to salvage a sizable number of castings which would otherwise be rejected. In this country the Army Air Forces allow only the helium arc process for this purpose, and this paper describes some experimental work on salvage which helped us out of some difficulties when we started to reclaim minor defects.

The welding process itself is relatively simple. A tungsten electrode is used in a concentric holder so arranged that a regulated stream of helium gas envelopes the arc. (Argon has also been used as a gas shield, but there are indications that the arc is not quite as stable as in helium. Most commercial argon has minute traces of water which probably contributes to this instability.) This inert gas shield protects the molten magnesium from oxidation, and avoids the use of flux of any kind. After the weld has been completed, a light film of white or brownish magnesium oxide covers the surface, and this is easily brushed off.

Extruded bare filler rod of the same composition as the casting (nominally 6% Al, 3% Zn, 0.2% Mn) is fed in separately. A conventional direct current welding machine, a tank of commercially pure helium with a flow-meter, and a preheating furnace controllable to $\pm 10^\circ$ F., complete the equipment. Reverse polarity is used, and the technique is similar to forehand gas welding. The arc is quite short and very steady,

but it tends to draw metal through the molten pool and deposit it behind the electrode. Welding must proceed smoothly and evenly, and the torch and filler rod should not be oscillated. To keep the filler rod from melting off in large globules it should be fed intermittently—or, preferably, at a slow rate from just outside the arc itself. A 30 to 45° angle should be maintained between electrode and plate, and a 90° angle between electrode and filler rod. To prevent gas entrapment, heavy deposits should be avoided since magnesium welds freeze and cool very rapidly.

Specifications usually limit the type of defect which can be welded. In general, cold shuts, cracks, and mis-runs cannot be welded. Some manufacturers limit the size of each weld to half the wall thickness in depth, and 1 sq.in. in area. This is somewhat severe, since holes entirely through some walls have been welded successfully. The number of welds per casting is usually limited by economics to four or five, but since nearly all of these castings must be solution heat treated after welding, there is actually small reason for limiting the number of welds from a metallurgical standpoint. For aircraft castings, it is also customary to designate critical areas in which it is necessary to radiograph each weld.

Aside from the conventional welding machine

variables, the temperature of preheating is the most important single factor. After a considerable amount of exploratory work, it was soon evident to us that one preheat temperature could not be used for all wall thicknesses. To investigate this problem, plates 6x6 in. were cast and then machined down to $\frac{1}{4}$ in., $\frac{1}{2}$ in. and 1 in. thick. The plates were radiographed, and to simulate the worst possible conditions the plates with the heaviest microshrinkage were chosen. Two 3-in. strips were then cut from each plate and three equally spaced semi-spherical cavities were cut into each strip. Each of these cavities, which played the part of a casting defect, was

The metallographic examination showed that the fusion was very good irrespective of preheat temperature. There was no difficulty in obtaining good fusion to either the sidewalls or to underlying passes. Normal plate has an average grain size of about 0.006 in., but the deposited weld metal had a much smaller grain size, in the neighborhood of 0.001 in. This deposited weld metal responded well to solution heat treatment and aging and, on pieces free from microshrinkage, it was quite easy to obtain welds which had 95 to 100% of the base metal strength. There was no area of eutectic melting in the base plate, and no heat affected zone of any kind next to the weld. It seemed that the concentrated heat source and the extremely rapid rate of cooling effectively prevented any adverse metallurgical effects in the adjoining metal.

Radiographs for some of the 0.25-in. specimens are shown opposite. All these welds were made in one pass, and carefully cleaned free from any heavy oxide deposits before they were preheated. Preheated at 700° F. each weld contained a large crack

Conditions for the Helium Arc Welding of Magnesium Castings

THICKNESS	OPEN CIRCUIT VOLTAGE	WELDING VOLTAGE	HELIUM FLOW, PER MIN.	ELECTRODE DIAMETER	FILLER ROD	POLARITY	PASSES
$\frac{1}{4}$ in.	65	17	61.	$\frac{1}{8}$ in.	$\frac{1}{16}$ in.	Reverse	1
$\frac{1}{2}$ in.	65	18	8	$\frac{1}{8}$	$\frac{1}{16}$	Reverse	1
1 in.	65	18	8	$\frac{1}{8}$	$\frac{3}{32}$	Reverse	4

NOTES: 1. Each cavity was one half of the plate thickness in depth, and equal to the plate thickness in diameter.

2. Where more than one pass was used the plate was reheated between passes.

3. The welding time for each cavity was approximately 10 sec.

for the $\frac{1}{4}$ -in. plate, 40 sec. for the $\frac{1}{2}$ -in. plate, and 40 sec. for each pass in the 1-in.

4. Each plate was 3x6 in., machined to indicated thickness, and had three equi-spaced welds.

5. Each plate was cooled in still air after the three welds had been completed.

half the plate thickness in depth and had a diameter which was equal to the plate thickness.

A series of strips was welded at preheat temperatures which varied from 100 to 700° F. in steps of 100°. In addition, several sets of plates were welded at room temperature (about 75° F.). The most satisfactory welding conditions other than preheat were first determined for each plate size, and these conditions were kept constant as the preheat temperature was varied. These welding conditions are listed in the table above, and represent our best judgment on the basis of experiments with both test plates and production castings. Each strip was preheated for at least one hour before welding and for at least 20 min. after one weld was completed and before starting the next. (The temperatures listed in this discussion are all surface temperatures determined with a chromel-alumel thermocouple.)

After welding, the excess metal was machined from the surface and the strips were radiographed under conditions capable of detecting a flaw of 2% of the metal thickness. The welds were also sectioned and examined metallographically.

directly across or around the weld. These cracks began to disappear at 400°, where two out of three welds were cracked. Preheated at 300° the cracks were gone, but the welds contained some porosity. As the preheat temperature was further lowered, the cracks did not reappear, but the amount of porosity progressively increased. Surprisingly, even at room temperature no cracks were present.

With the 0.50-in. plate no cracks were developed at any preheat temperature. Porosity was absent after 700 and 600° preheat, and these welds were very satisfactory. At 500° one weld out of three was slightly porous, and this porosity increased gradually as the preheating temperature dropped, so that at room temperature there was an excessive amount of porosity. With the 1-in. plate exactly the same results were obtained.

Apparently, a minimum amount of metal must be present to overcome the cooling stresses if cracking is to be prevented. In the thin plates the high temperatures promoted low hot strength, and at these temperatures there was insufficient strength to resist cracking. With thicker plates or lower preheat temperatures, more strength

was available or developed, and the cracks were eliminated. Porosity, on the other hand, was less prevalent as the preheating temperature increased, and it was evident that the weld puddle had to remain molten long enough to allow the entrapped gas to escape. For this reason, the thicker the plate, the higher the preheat temperature.

Individual technique influenced the cracking and porosity to some extent, but the above results proved generally applicable to production castings. Therefore for all salvage work the following preheat conditions were maintained:

1. On all walls 0.25 in. and under the work was to be preheated to 300° F.
2. For all other thicknesses the preheating temperature was 640° F.

A temperature of 640° F. was chosen for the heavier work because there is danger of eutectic melting above 680° F. If a temperature of 700° had been specified it would have been necessary to heat the material in the as-cast condition at a rate slow enough to allow the Mg-Al-Zn eutectic, which melts at about 680° F., to go into solid solution and diffuse into the matrix. At 640° these precautions were unnecessary, and this temperature was adopted for the heavier wall thicknesses.

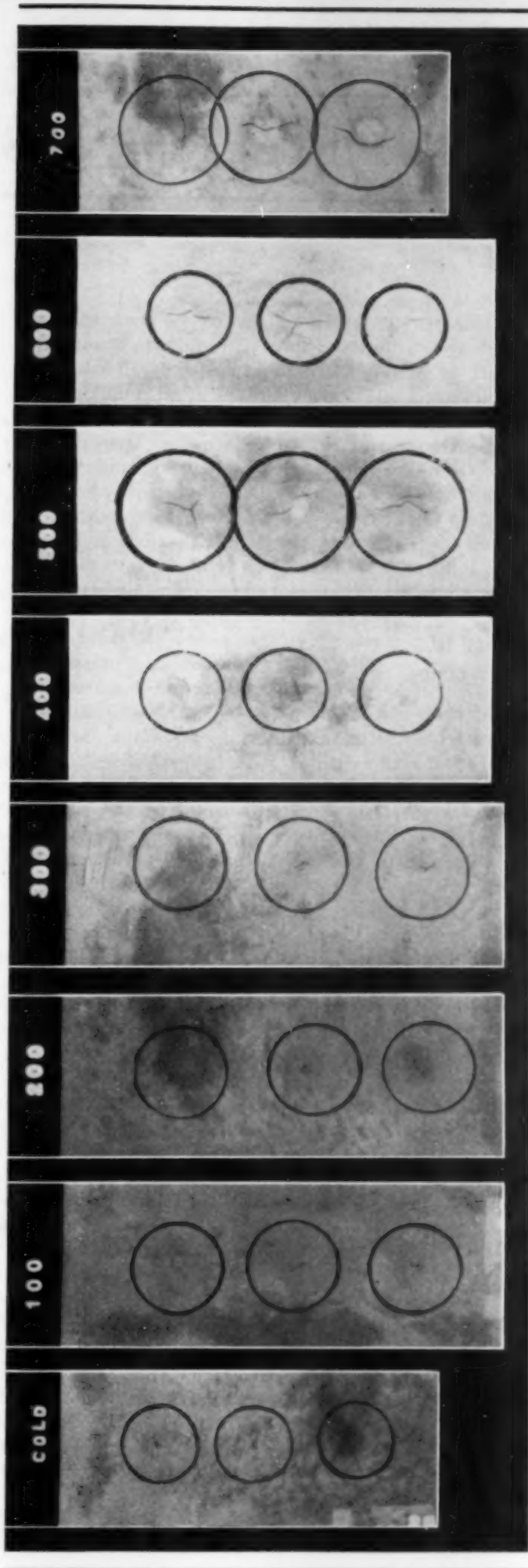
If the preheating conditions are watched carefully, 85 to 90% of the welds in walls ½ in. and over should be acceptable. For thin walls (¼ in. and under) 75% of the castings should be salvaged, and the rest lost from porosity and a few cracks.

In summarizing this work it should be pointed out that the number of test samples was small, but that the results were, in general, verified on production castings. We have found that:

1. The helium arc process is capable of salvaging a large number of magnesium castings.

2. For walls ¼ in. and under a preheating temperature of 300° F. should be used. For walls ½ in. and over a temperature of 600 to 700° F. should be used, with 640° recommended for safety.

Radiographs, Reduced to About Half Size, of Welds Made in ¼-In. Cast Strips, Show That Cracks Exist at Preheat Temperatures of 400° F. and Above, Whereas Porosity Increases as Temperatures Drop Below 300° F.



CRITICAL POINTS

By the Editor

CONSIDER this phenomenon in our American "capitalistic" economy: A multitude of men collect scrap and sell it to a large company that has invested millions of dollars in steel-making plant and enormous continuous mills for rolling strip-sheet. Spaniards and Portuguese strip the bark from the cork oak, ship it here, and we search our Southern forests for regions where the tree can thrive. An American firm, whose founder, CHARLES McMANUS, is still alive, re-invests its profits, millions upon millions, in rolling mills and annealing furnaces to make thin strip,

**Cost: 10
for a cent**

to clean and lacquer and tin it, to engrave and print the metal sheets, and for hundreds of presses to stamp out the round disks and crimp their edges. More money, brains and engineering ability for machinery to grind the cork, re-press it into extruded rods, bake it to set the binder, and slice it into round seals about the size of a quarter. Finally, to install in one long wing of this amazing factory a battery of machines, each one with hopper-loads of metal and cork disks, whose duty it is to cement the cork into the metal cup and deliver a completed bottle-cap in a literal torrent—a million caps produced in the few minutes needed to stroll through the room, marveling at the ingenuity that culminates in making this tiny device we pull off the top of a beer bottle and throw away without thinking. Their cost must be enormous—but you're wrong. They sell ten for a penny!

CROWN CORK AND SEAL CO.'s management has a passion for independence, as shown by the efforts to develop an American source of cork—the tree thrives only around the Mediterranean—and by its installations for converting hot rolled sheet into strip of requisite thickness, temper and finish. A five-stand cold mill reduces hot mill coils, 0.075 in. thick, down to 0.010 in. Next is a bright annealer of unique design, where the strip loops up and down eight times in a square tower 45 ft. high, each pass in its own compartment, electrically heated, and finally reaches 1450° F. Then come more loops between water-cooled walls; all is in controlled atmosphere; passage time is about 90 sec. and the strip

emerges, not dead soft, but bright annealed and of correct hardness so that a final pinch pass to planish the mossy surface will give it the proper spring qualities of a bottle cap. Electrolytic tinning lines are similarly built-on-end, as it were, and consist of deep tanks into which the strip is looped over top and bottom rollers. The tinning cell itself has dozens of slab-like anodes,

**Bottle caps
and bottling
machinery**

30 in. wide by 7½ ft. long, hung vertically, and the strip threads up and down, up and down, past each one in turn at the rate of 625 ft. per min. Final "flowing" of the plate is a flash melting between vertical banks, each one comprising a multitude of small gas burners. . . . All these were explained by WALTER McMANUS, the firm's assistant treasurer, and GEORGE NAUSS, its genial metallurgist. The latter's principal activities are in connection with the company's own highly specialized machinery for making the various bottle and jar seals, in dozens of shapes and designs, and in the shops for producing all manner of bottling machinery. The dairy equipment is constructed of the stainless steels almost exclusively, they having demonstrated their ultimate economy in a sanitary way. . . . Oh yes—an example of metal replacing cardboard: A new sanitary cap for milk and cream bottles consists of white lacquered sheet steel, 0.003 in. thick, glued to a thin paper lining. It's considerably stiffer than cardboard and only about one-third as thick.

SOMEWHAT EMBARRASSED by registering ignorance when the word "fractography" was used by CARL ZAPFFE, my host at Rustless Iron and Steel Corp.'s research division in Baltimore. It implied a lack of editorial curiosity, since the word has already been used in these pages, but was rewarded, instead of being punished for my sins, with a first-hand account of an interesting new technique. Preparation of a sample for micro-examination is simplicity and rapidity itself. The specimen is merely broken, stuck into

a mounting stage with plasticene, and one of the facets is brought level with the aid of binoculars, and this face is then ready for examination under high power, for the tiny crystal breaks along a plane flat enough for 750 to 2000 \times or so. Obviously such a surface is undisturbed by the cold work of polishing or

the chemical attack of etchants; being a plane of weakness it quite likely is the most informative win-

dow of observation to a searcher for defects. Aside from its use on some fundamental studies of sub-crystalline architecture, fractography is of day-by-day use in examining high alloys near the borderline where undesirable (or desirable) micro-constituents exist. Some of the latter are unerringly located, even more readily than by looking at plane-polished specimens. As to the latter, much use in this laboratory is made of "tempering" (mild heat in oxidizing atmospheres) to color the various metallic phases, rather than etching them. . . . Not to be caught napping, similarly, about another new idea in microscopic study, read with interest the published text of CHARLES BARRETT's lecture, prepared for the February meeting of the Institute of Metals Division, A.I.M.E., but undelivered because the meeting was cancelled—war in Europe and Asia, and all that. He places a photographic plate with fine emulsion close to a smooth metal surface and grazes the latter with a beam of X-rays; reflections from disturbed places in the crystal lattice are registered on the plate, which is then developed and studied under a microscope at 100 diameters. This method of mapping the distribution of strains at and near the surface obviously rests on the ability to prepare a smooth surface free

X-ray microscopy

from foreign strains placed there during polishing, and this is fortunately possible by a succession of picklings and electrolytic etchings. . . . Notable among the early indications of this new X-ray microscopy is the conclusion that cold work produces slip bands wherein are local bendings or rotations of the crystal lattice of less than 0.1° ; the disturbed metal at and alongside the slip bands is in a layer no thicker than 0.003 mm., and that there is no strain in the intervening metal between slip

NOMINATIONS FOR NATIONAL OFFICERS

In accordance with the Constitution of the American Society for Metals, the Nominating Committee appointed by President KENT R. VAN HORN met on May 24 in Chicago. The Committee consists of WALTER CRAFTS, chairman; R. M. BRICK, C. E. CARLSON, Lt. Col. PAUL C. CUNNICK, HOWARD E. HANDY, WALTER A. HILDORF, H. K. IHRIG, A. J. SMITH, and L. C. WHITNEY. By unanimous action of the Committee the following nominations were made for A.S.M. national officers:

For President (One Year)

CHARLES H. HERTY, Jr., assistant to vice-president, Bethlehem Steel Co., Bethlehem, Pa.

For Vice-President (One Year)

A. L. BOEGEHOLD, head, metallurgy department, Research Laboratories Division, General Motors Corp., Detroit.

For Treasurer (Two Years)

HAROLD K. WORK, manager of research and development, Jones & Laughlin Steel Corp., Pittsburgh.

For Trustees (Two Years Each)

WALTER E. JOMINY, chief metallurgist, Dodge Chicago Plant, Chrysler Corp., Chicago.

JOHN CHIPMAN, professor of metallurgy, Massachusetts Institute of Technology, Cambridge.

The Constitution of the Society provides that additional nominations may be made prior to July 15 by written communications addressed to the Secretary of the Society and signed by any 50 members. If no such additional nominations are received prior to July 15, nominations shall be closed and the Secretary, at the next annual meeting in the fall, shall cast the unanimous vote of all members for the above candidates.

bands. Such a cold worked structure persists during annealing until entirely devoured by completely recrystallized grains. Cast metals are also shown to be made of highly imperfect crystals, full of "micro strain peaks" as close together as 0.02 mm. Lastly, age hardening is accompanied by complete reorganization in the crystals of solid solution, evidenced by the distortion or rotation of fragments 0.003 to 0.005 mm. in size. Thus, age hardening is similar to the mechanism of hardening by cold work—a finding that lends no support to the older theory that age hardening results from a critical dispersion of precipitated particles which "key" the slip planes.

FOUND that much work had been done, in

Rustless's Research Division on the "colors" of stainless alloys — "color" in the sense used by jewelers and metal artists. ALEX FEILD, director of the division, said they started with an optical analysis of the light reflected from stainless steels. What change in analysis or treatment is responsible for the various appearances ranging from the cold blue of steel to the warm yellowish-white of silver? During their attempts to match colors of other metals used in

Advances in the stainless steels

the arts, a black, non-reflective surface was discovered; it is a good camouflage of war matériel. The stainless steel is merely held 15 to 20 min. in molten sodium dichromate at 740° F., cooled and washed. Corrosion resistance is not impaired and the "color" is so hard and tight that blackened sheet can be formed. . . . Found that KENNETH BLOOM, assistant director, had been giving considerable attention to the martensitic 16-2 chromium-nickel steel described by his associate STANLEY WATKINS in *Metal Progress* for July 1943, and among other things has developed in it really excellent spring qualities: 225 to 350,000 psi. ultimate strength, 180 to 300,000 psi. proportional limit, 11.6×10^6 psi. modulus in torsion. . . . Finally, spent several hours in the electric furnace department watching the pyrometric work. Messrs. WEITZENKORN and KLINGEL use a short platinum thermocouple, properly protected by a fused silica tube at the end of a long pipe handle, for getting *correct* temperature of the bath when it is supposed to be ready to pour. This single accurate reading, more reliable than an optical pyrometer obscured by smoke or slag, has been of such importance in improving ingot quality that the next step is some device that will record the temperature of the steel *continuously* during the heat. Not too impossible.

COMBINED metallurgy with aesthetics while visiting the brass cartridge case line of Rheem Mfg. Co. at Birmingham, where — in keeping with a summer temperature — the predominantly feminine staff kept cool in shorts. From an *engineering* standpoint, the plant is notable for its complete conveyor systems, possibly dictated by the sex of the workers, for 90-mm. cases weigh 11 lb. each, about as much as a woman can handle at the rate of 600 per hr., day in and day out. Consequently, transit of work from stage to stage — especially through process annealing furnaces and pickling equipment — is in baskets made of stiff stainless steel wire, wherein 60 cases are racked, open mouth downward, so pickling

and wash solutions sprayed into them may readily drain out. The automatic pickling machine, by the way, is made of stainless steel complete — roller hearth, hoods, sump, nozzles, and so on. . . . HARRY WILMOT, director of the metallurgical laboratory, said that this line in the Birmingham plant was started with a complete set of punches

Relation of die design to auxiliary operations

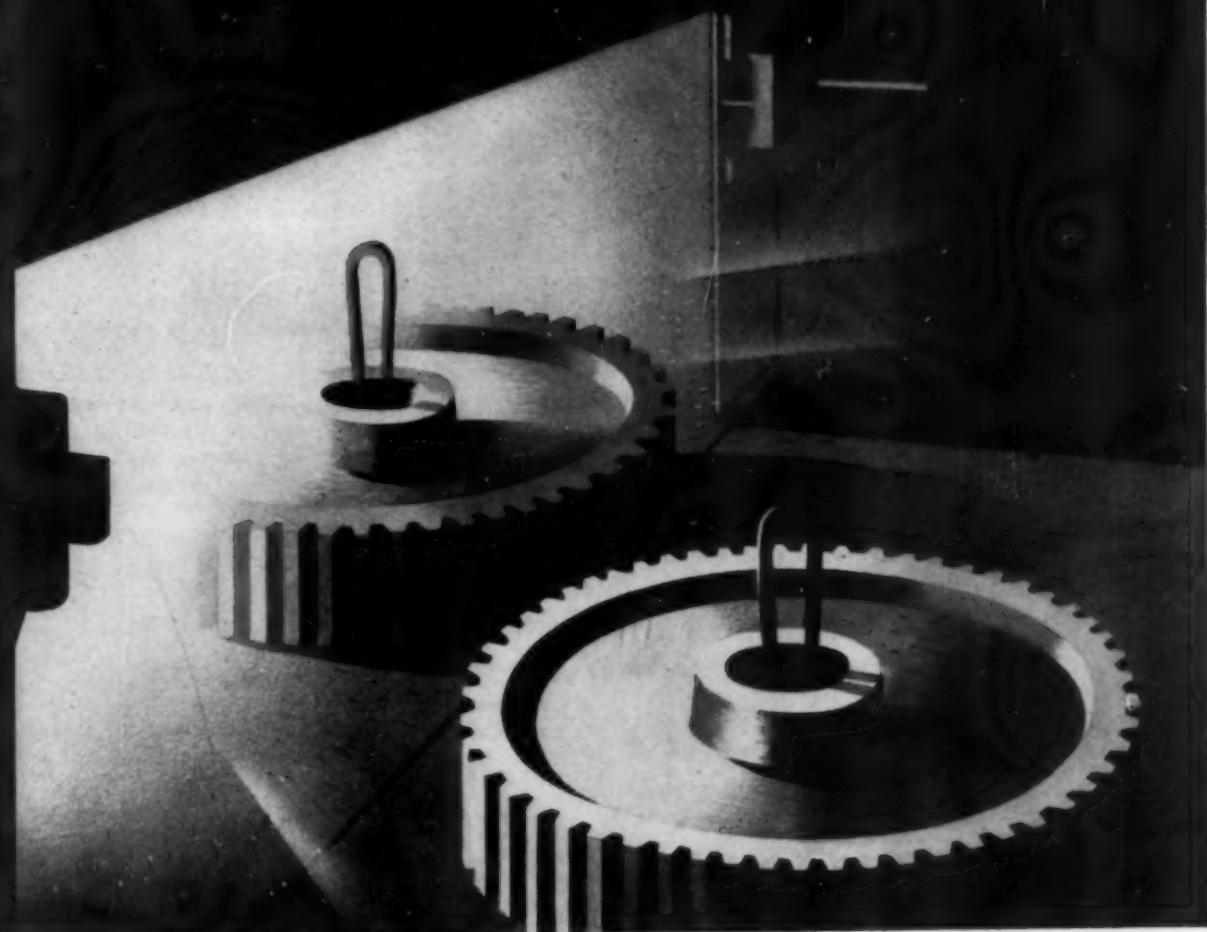
and dies from Rheem's Chicago plant, where they certainly ought to know how to make cartridge cases, having produced (on the side, more or less) the enormous 152-mm. cases that stand 4½ ft. high for the French battleship Richelieu's secondary battery. However, some changes from the Chicago practice were immediately necessary; for example, at Chicago the annealing is done in open flame gas furnaces; at Birmingham in open-ended electric furnaces. Differences in oxide formation and loss in cleaning were enough to require slight modifications in punches and dies. . . . Another heating quirk: Radiant heating for mouth anneal proved erratic because the smooth brass surfaces *reflect* the heat waves rather than absorb them. . . . These large 90-mm. cases are necked considerably, down to projectile caliber, and this, together with the tapering operation, is critical — much deformation is required in a thin-walled object with no inside support. At Birmingham this operation is done mouth up, and WILMOT said it involved a peculiar problem in die lubrication. The grease has to be stiff enough to stick on the vertical walls of the case without draining down, yet if there were too much it would be trapped between the forming die and the case wall, and actually dent the case. Grease is therefore spread on by hand, as it is in the heading operation as well, for in the latter a little extra grease, if present, will produce a case with a base too thin.

MAY I TAKE a few moments to vent my contempt for those "research organizations" that offer "technical service for building tomorrow" or such-like poesy, staffed by "researchers" who don't know, to judge from the letters they write, that *The Engineer* represents the name of a leading British publication. A

Current gripe

constant stream of requests comes from them to the editorial desk, each one asking and always thanking in advance for engineering information which, if available, would have too much market value to give away. The letters go direct into the waste basket, but evidently the blind is leading the blind, and God help reconversion and the bright new tomorrow built on their advice.

Molybdenum steels require relatively high tempering temperatures and therefore are relatively free from internal stresses.



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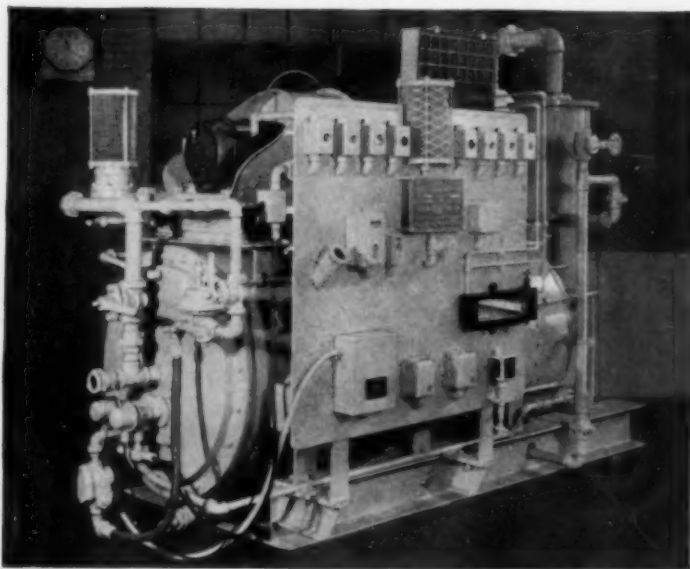
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PERSONALS

O. J. HORGER, formerly in charge of railway engineering and research, has been appointed chief engineer of the railway division of the Timken Roller Bearing Co. C. L. EASTBURG, in charge of the design of bearings and parts as applied to locomotives, has been appointed assistant chief engineer of the rail-

way division. P. C. PATERSON, active in inspection and procurement of material and in equipment applications, will be service manager of the railway division.

TRACY V. BUCKWALTER, chief engineer and vice-president of the Timken Roller Bearing Co., has retired under the company's retirement annuity plan, but will continue to serve in a consulting capacity. ALBERT L. BERGSTROM, formerly executive engineer, has been elected vice-president of all engineering.



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GUNTHER MOHLING has been presented with a cash award and certificate of appreciation by the board of directors of the Allegheny Ludlum Steel Corp., Pittsburgh, for his contribution to the development of a series of high temperature alloys, one of which is particularly suited to jet propulsion aircraft.

ROY D. HAWORTH, Jr. has resigned as metallurgist with Wymann Gordon Co., Harvey, Ill., to become chief metallurgist with Lehigh Foundries, Easton, Pa.

ROYDEN J. WHEELER, formerly chief metallurgist, Riverside Metal Co., and engineer of tests at the Richmond plant of the American Locomotive Co., is now chief metallurgist, A. G. Carter, Inc., Gastonia, N. C.

Promoted by Wright Aeronautical Corp.: JAMES R. COXEY from assistant metallurgist to metallurgist.

WILLIAM L. O'BRIEN, formerly district sales manager in Indianapolis for Jessop Steel Co., is now manager of the stainless steel division at Washington, Pa.

Transferred by Aluminum Co. of America: W. M. MUELLER from New Kensington, Pa., to the Phoenix, Ariz., works.

Transferred by National Tube Co.: WALTER K. GRAHAM from Dayton, Ohio, to Dallas, Texas, as sales engineer for the tubing specialties division.

LOUIS SILVERMAN, formerly superintendent of inspection, Bowers Products Corp., Ecorse, Mich., is now metallurgist, Radio Corp. of America, RCA Victor Division, Harrison, N. J.

ARTHUR G. THAYER has resigned as metallurgist in the research laboratory of Carnegie-Illinois Steel Corp., and is now engineer for National Metal Products Co., Pittsburgh.

HENRY S. FREYNIK has been promoted to chief metallurgist for the Riverside Metal Co., Riverside, N. J.

ERNEST H. WYCHE has been made metallurgical engineer for the Titanium Alloy Mfg. Co., Niagara Falls, N. Y.

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PERSONALS

MAXWELL GENSAMER ☼, formerly professor of metallurgy and member of the staff of the Metals Research Laboratory, Carnegie Institute of Technology, has been appointed professor of metallurgy and head of mineral technology at the Pennsylvania State College.

M. T. ARCHER ☼ of the National Supply Co., Torrance, Calif., has joined a W.P.B. group of technical advisors to the Chinese Government, stationed at Chungking.

ROBERT SERGESON ☼, formerly with Crucible Steel Co. of America, has been appointed chief metallurgical engineer of the Rotary Electric Steel Co., Detroit, succeeding L. L. FERRALL ☼, resigned.

LEE A. DAINES ☼, formerly district manager in Chicago for Hepenstall Co., Pittsburgh, has been elected vice-president and general manager of sales for the company in Pittsburgh.

NORMAN I. STOTZ ☼ has returned to Braeburn Alloy Steel Corp., Braeburn, Pa., as vice-president in charge of sales, after ten years' absence when he was with Universal-Cyclops Steel Corp. as metallurgist.

Appointed manager of sales for Lindberg Engineering Co., Chicago: NORBERT K. KOEBEL ☼, who will also continue to serve as director of research. ROBERT W. DOUGHERTY ☼, formerly of the Chicago office, has been transferred to the Lindberg Detroit sales office.

JOHN R. QUINZIO ☼, formerly with the Buffalo Foundry & Machine Co., is now connected with the Carborundum Co. of Niagara Falls as maintenance engineer.

W. R. TOEPLITZ ☼ has been made vice-president in charge of engineering research for the Bound Brook Oil-Less Bearing Co., Bound Brook, N. J.

PVT. ORLANDO C. METCALF, JR. ☼ has been assigned to the Frankford Arsenal, Philadelphia, in the research laboratories, carrying an army classification of metallurgist.

JOHN A. COMSTOCK ☼, formerly engineering metallurgist with the United Aircraft Corp., East Hartford, Conn., has been appointed director of research and metallurgy for all divisions of H. K. Porter Co., Inc., and will be in charge of the central research and testing laboratory located in Pittsburgh.

GEORGE A. PETERSON ☼ and GEORGE H. CAMPBELL ☼, both of Pratt & Inman, Worcester, Mass., along with THEODORE C. CEDERHOLM ☼, formerly district sales manager for Wyckoff Drawn Steel Co., and JOHN R. VAN NEST, formerly sales engineer with American Brass Co., have bought the firm of Pratt & Inman and will continue to operate it under this name.

F. E. BENSON ☼ has been appointed sales engineer for American welding products in Canada, and is connected with the Canadian Ramapo Iron Works, Ltd., Niagara Falls, Ont.

D. F. MCFARLAND ☼ has announced his retirement this summer as professor and head of the department of metallurgy, Pennsylvania State College.

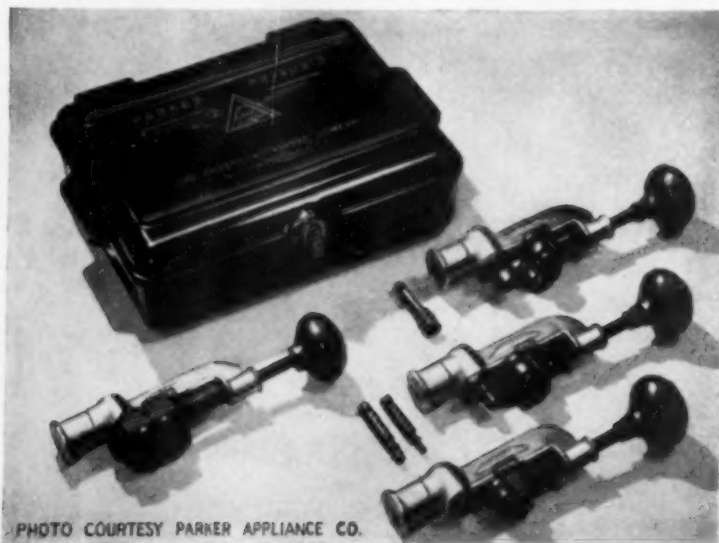


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Q. Which x-ray film for shortest possible exposures of thick welds?

A. Kodak's Type F

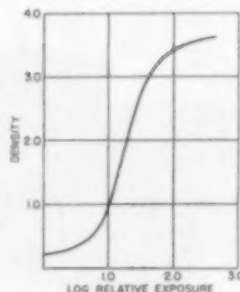
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tant requisite is "shortest possible exposure." Eastman Kodak Company, X-ray Division, Rochester 4, N. Y.

Characteristic Curve, Kodak Industrial X-ray Film, Type F, with double intensifying screens of calcium tungstate.

Development: Five minutes, at 68° F., in Kodak X-ray Developer or Kodak Liquid X-ray Developer.



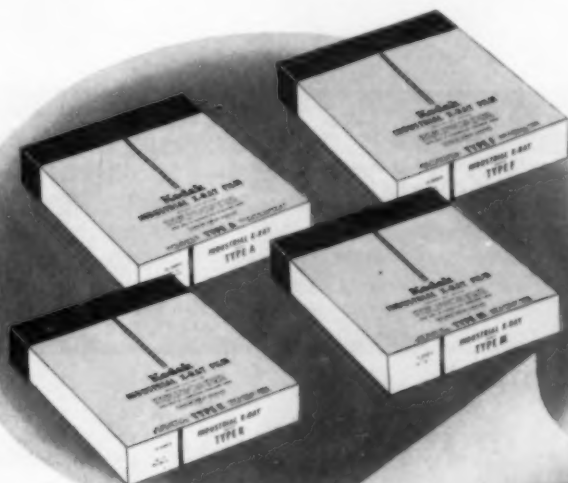
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PERSONALS

DONALD G. CLARK ☉ has retired as a member of the board of directors of Firth-Sterling Steel Co., Pittsburgh, but will remain in an executive and consulting capacity.

C. R. AUSTIN ☉, professor of metallurgy at the Pennsylvania State College, has accepted an appointment as assistant to the president of the Meehanite Metal Corp., New Rochelle, N. Y.

ELMER A. TERWELL ☉, formerly with the Chicago office of the Driver-Harris Co., has been appointed assistant manager of the Chicago plant of Salkover Metal Processing.

Promoted by Jones & Laughlin Steel Corp.: L. C. BERKEY ☉, from district sales manager, St. Louis office, to district sales manager, Chicago office.

W. E. REMMERS ☉ has been elected vice-president of Electro Metallurgical Co., New York.

W. H. GRAVES ☉ has been promoted from chief metallurgist and quality control manager for Packard Motor Car Co., Detroit, to executive engineer.

GORHAM W. WOODS ☉, formerly process engineer of the Dickson Gun Plant of Hughes Tool Co., has joined the engineering staff of the Lincoln Electric Co., Cleveland.

LORIN L. FERRALL ☉ has been appointed metallurgical director for the Crucible Steel Co. of America, New York City.

LT. COL. A. E. R. PETERKA ☉, assistant chief of Readjustment Division, Air Technical Service Command, Wright Field, Dayton, Ohio, has been relieved of active duty with reversion to inactive status so that he may serve as consultant to the Defense Plant Corp., Washington, D. C.

ROY L. ANDERSON ☉, in addition to his regular position as metallurgist with the Star Brass Mfg. Co., Boston, has taken on the additional responsibility of consulting metallurgist with the W. J. Dunn Co., South Boston, Mass.

Promoted by Callite Tungsten Corp., Union City, N. J.: LAWRENCE HALLERAN ☉, formerly assistant supervisor of the Alloy Wire Mill, to sales manager of the Wire Division, succeeding HAROLD M. MALM ☉, who has resigned to join Little Falls Alloys, Inc., West Paterson, N. J., in an executive capacity.

RALPH HAMMERSLEY, Sr. ☉, formerly with the Frederick Hart Co., Poughkeepsie, N. Y., is now in charge of the Troy, N. Y., office of Crawford Engineering.

RAY D. McMULLIN ☉, formerly chief production engineer, Lawrence Aeronautical Corp., Linden, N. J., is now affiliated with the Kiekhaefer Corp., Cedarburg, Wis., as executive engineer.

Promoted by Electro Refractories & Alloys Corp., Buffalo: GEORGE B. MICHIE ☉, from metallurgist, in charge of purchasing and priorities, to vice-president in charge of sales.

CHARLES F. BOWER ☉, formerly with Republic Steel Corp., is now employed by Austenal Laboratories Inc., New York, as metallurgist.



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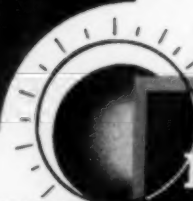
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